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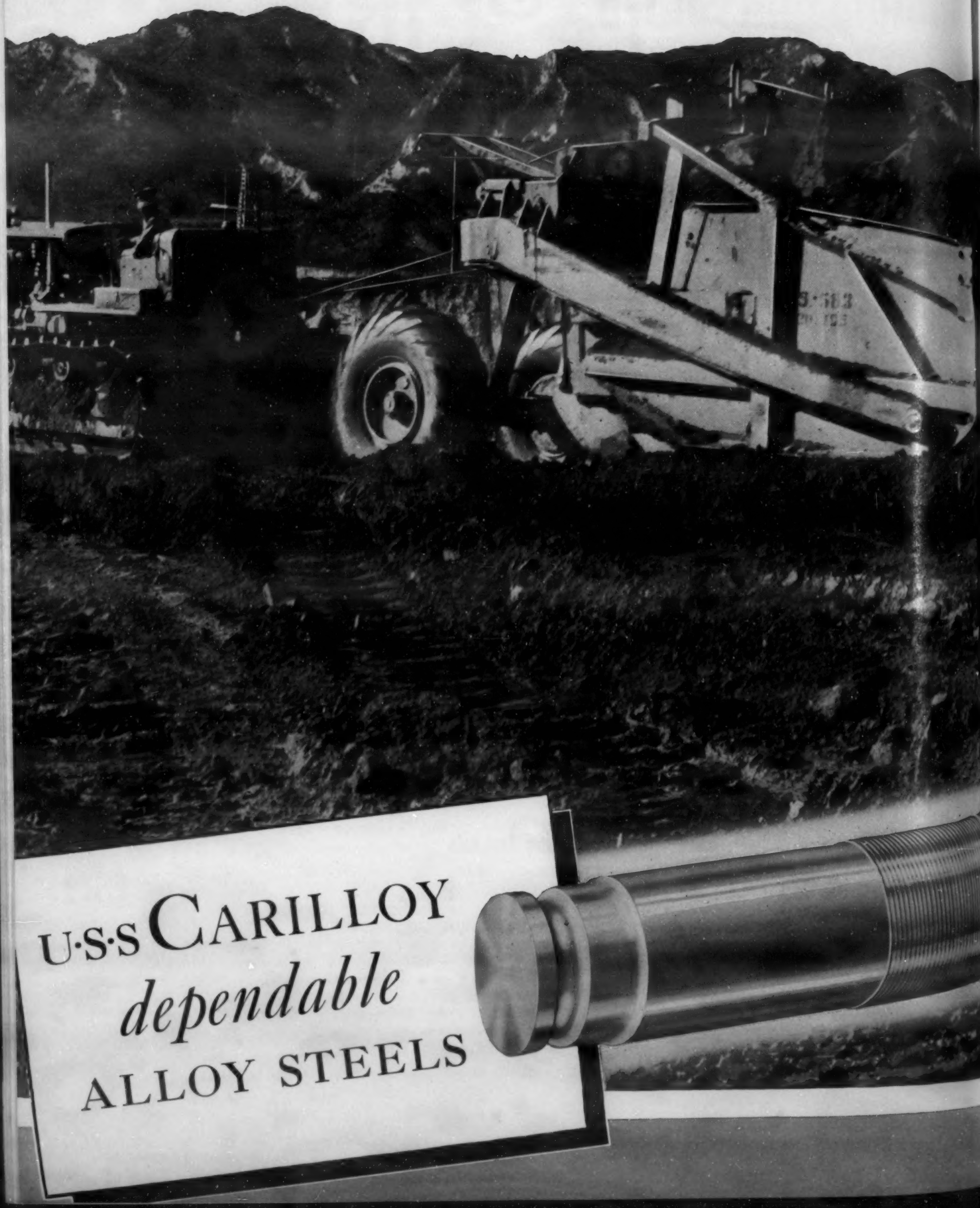
In the *Vital Parts*



U·S·S CARILLOY
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Don't Shoot the Organist

The editor of our British contemporary, *Metal Treatment,* in the spring issue, casts his eye over the accumulating mass of metallurgical literature and is not wholly pleased with what he sees. He finds a great deal of duplication, in that many investigations merely confirm what might have been justly inferred from what is already known, had the investigator become sufficiently steeped in the literature. To facilitate acquaintance with what is known so far, he advocates publication of periodic critical surveys, i.e. what we call "correlated abstracts."

But beyond the case of one who wastes time through ignorance of what others have already found out, he bewails the tendency of investigators to work in a rut, to stick to the same old line of study, developing more minutiae that none but a very few specialists care to know and urges that the investigator seek to find out "something there is good reason for wanting to know," something that will make lots of metallurgists rise up and call him blessed.

We can sympathize with this feeling, for in reading galley proof of abstracts we often pause to wonder, "So what?", when we cannot see why a man should have spent good time and money in doing what he did.

In the last analysis, however, the situation is hardly as bad as it looks. The commercial director of research always asks whether there is "good reason for wanting to know" before approving a suggested project, and it is one of his more important functions to stop it if it does not have a good chance of reaching that goal. Commercial investigations that get nowhere or are repetitious are seldom offered for publication unless they are merely aimed as propaganda. The "so what?" type are more likely to be academic researches. That these tend to run in ruts is not solely due to laziness or provincialism of the professor, but to the fact that striking out into another field would require apparatus that the college cannot finance; if the student is to do any research, he must do it with what is at hand. He needs some introduction to research and the work is done not primarily for the information it gives the world, but for the training it gives the student.

The editor of *Nature,* discussed, in the April 29 issue, the establishment of an aeronautical research institute in Australia, against which it had been urged that it would be better to concentrate such research in England and let the colonies utilize the data. The editor pertinently says "In the first place, no one can appreciate fully the results of scientific research unless they themselves are engaged in similar work. It does not matter in the least that Australia may not be in a position for many years to contribute anything of first class importance to the science of aeronautics. If no start is made, there never will be such contributions."

From this point of view, the real peeve is not that dinky little investigations are made for purposes of instruction, but rather that they are too often written in lengthy fashion and phrased as though they were all-important, instead of being merely a brief note to record what trace of new information was found out.

Up to the point where repetition becomes pointless, corroborative data are worth recording in order to clinch a fact. Information only of value now to a few specialists may be of wider interest a little later on, if those specialists are engaged in unearthing something that the world will finally wake up to find it ought to know more about. So even little patches of apparently useless information should be published and made findable by abstracting or listing, for they may save many people from duplicating the work if and when there is a real need for the data. Glancing back at the first issue of METALS AND ALLOYS, we note an abstract on the use of spectrographic metallurgical analysis. Ten years ago this article was of interest only to a "little group of specialists." Today the method is accepted routine, and the early, fragmentary, articles helped to make it so.

It would be nice if every professor and every research director had second sight and could select only those projects that will bring forth facts that are both new and needed. Actually, anyone who can score one real hit in ten tries is doing pretty well. Publication of the by-products, or the negative information, of the other nine, should be helpful to others who might be tempted to deal with the same

(*Continued on page 258*)

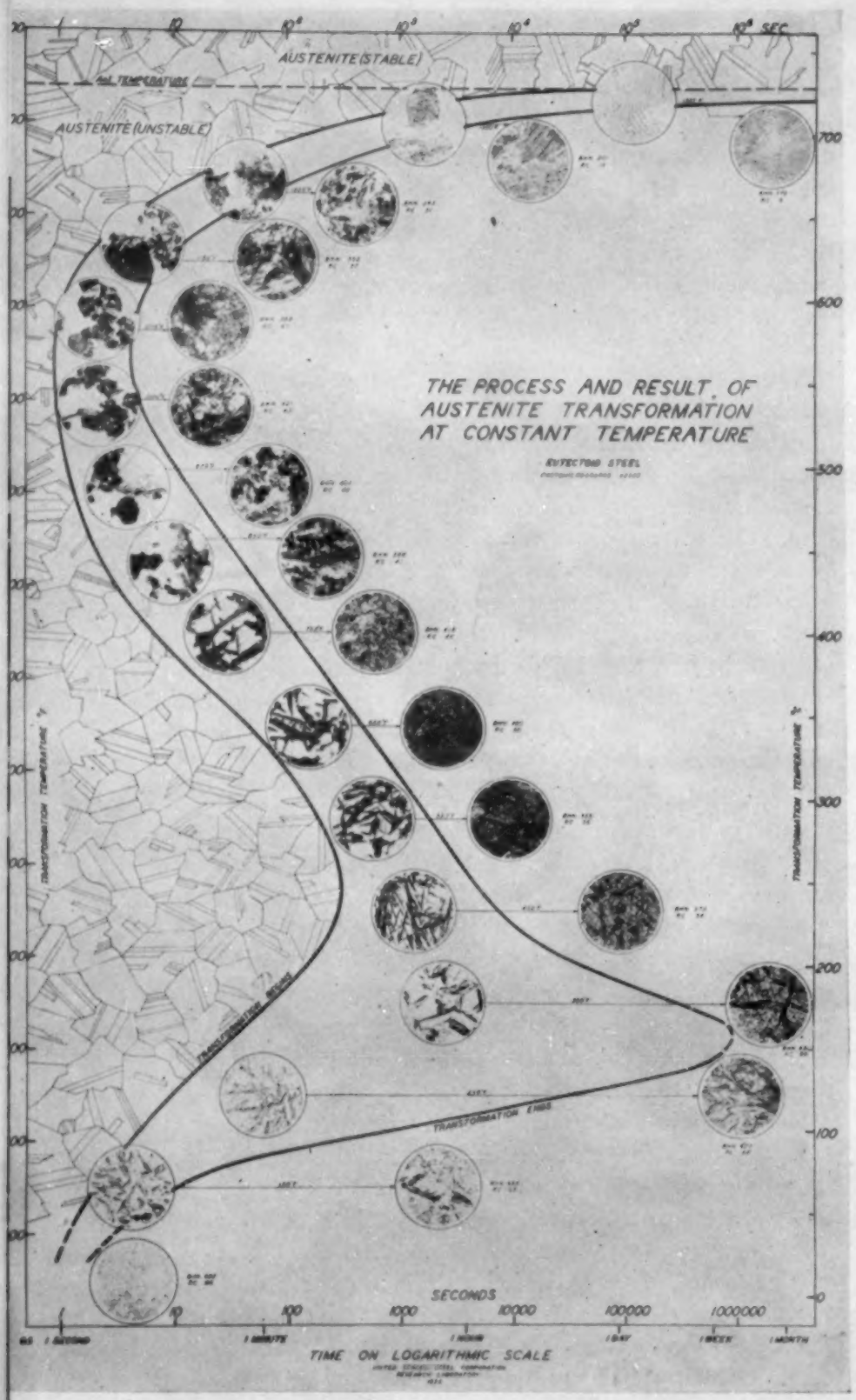


Fig. 1. Reproduction, on a Very Small Scale, of the U. S. Steel Corp. Research Laboratory's S-Curve for Eutectoid Steel.

The Industrial Application of AUSTEMPERING

by E. E. LEGGE

Director of Research, Worcester District, American Steel & Wire Co., Worcester, Mass.

Ever since this magazine, in January 1937, published the first practical discussion of the brilliantly conceived "S-curve" by the group responsible for its development, metallurgical interest in the austempering process of heat treating steel—and particularly in its industrial application—has steadily mounted. This development, one of the most fundamental changes in steel treating practice in several decades, has passed well beyond the laboratory stage, yet the details of its industrial-size application have not heretofore been published. In this article, Mr. Legge describes such an application of the process at the Worcester plant of his company. The unique results achieved so far in treating certain products, the large scale to which the process is evidently adaptable and, most important, the probable future position of austempering in the general heat-treating picture are clearly portrayed.

Many of our readers probably have the large S-curve published in our January, 1937, issue tacked up on their office walls for ready reference. The present article, which contains so much heretofore unpublished practical operating data on the industrial process, will very nicely supplement the wall-chart as reference material.—The Editors.

AUSTEMPERING IS THE METHOD of treating steel which comprises heating it until it is substantially austenitic, quenching it into a heat-abstracting bath at some selected temperature between about 350 and 800 deg. F., holding the steel at a chosen temperature within the range until transformation is substantially complete and then cooling to room temperature.

The classic work by Davenport and Bain[1] of the United States Steel Corp. Research Laboratory, Kearny, N. J., on the constant-temperature transformation of austenite is the foundation of the austempering process. One of the results of their study was the discovery of a new transformation product (bainite) offering possibilities of combined hardness and ductility not previously attainable by the conventional "quench and temper" method of heat treatment. From this observation they proceeded naturally to the development of a heat-treating method[2] to produce this structure. The first discussion in the periodical literature of the practical phases of the process, together with a complete detailed "S-curve" for eutectoid steel, appeared early in 1937[3] and was followed very shortly by additional information[4] on the possibilities and limitations of austempering as a heat-treating method.

This S-curve for eutectoid plain carbon steel is reproduced on a much smaller scale than the original in Fig. 1. Essentially it is a temperature-time diagram, with a logarithmic time scale to accommodate the tremendous range of time intervals (1 sec. to 1,000,000 sec.) embraced by the diagram. What it shows, briefly, is that steel made austenitic and then held at a constant-temperature level below the A_{e1} point will require for the beginning and for the completion of its transformation definite time periods that are characteristic of the heated steel and of the constant-temperature level at which it is held. The zone to the left of the left hand curve represents the area of completely untransformed, but unstable, austenite; the distance between the curves indicates for each constant-temperature level the time required for transformation to go to completion (or, in the lowest temperatures as far as it appears to go) once it has begun. The structures obtained by transforming at various temperature levels range from coarse pearlite near the top of the curve through fine pearlite and bainite down to martensite at the bottom.

The Process and Product

The austempering process is, in practice, confined to those constant-temperature levels that produce all or mostly all bainite in the transformed steel—temperatures usually between 350 and 750 deg. F. Structurally, bainite is a feathery, acicular constituent, *somewhat* similar to slightly tempered martensite in appearance, but generally clearly distinguishable from it by an experienced metallographer with adequate equipment. Mechanically, austempered steels consisting largely of bainite are remarkably tougher and more ductile at hardnesses above Rockwell C 43 than are similar steels quenched and tempered to the same hardnesses. Austempering, therefore, lends itself admirably to the treatment of certain types of spring products and other hardened parts requiring high elasticity, some of which, so treated, were shown for the first time at the National Metal Exposition in Detroit, October 1938. Fig. 5, a photograph taken on that occasion, shows some typical applications for which austempering is ideally suited.

Probably the simplest way to depict the fundamental differences between the conventional quench and temper method of heat treating and the austempering process is by comparative diagrams like Figs. 2 and 3. These show very clearly the manner in which austempering avoids the martensite stage in heat treating, by permitting the steel to transform *directly* at temperatures selected to give the properties desired.

Small Scale Operation

These diagrams suggest that the austempering process need be no more complicated than the conventional quench and temper method of heat treatment. That this is actually the case with respect to the equipment concerned is indicated by Fig. 6, which is a schematic diagram of the major units employed in the process; these are exactly the same as those often used in the customary quench and temper treatment, except that the order of their use is different. Fig. 7 shows the austempering equipment as set-up in the United States Steel Corp. laboratory at Kearny for austempering small lots experimentally. The hardening furnace at the right, the small hot salt bath in the middle, and the water container at the left are actually all that are required for small-scale application of the process.

Industrial Large-Scale Austempering

Before the details of the equipment operating at Worcester are presented, it may be well to outline why equipment of that particular type was selected, and particularly, why it was decided to austemper finished *parts*, in batch, rather than *wire* continuously.

Because of the long time of transformation required (See Fig. 1) for all but relatively low austempered hardnesses, a continuous process for austempering wire at economical speeds would require unreasonably long lengths of equipment, possibly running up to 500 or 600 ft. Furthermore, as tests on an experimentally-operated continuous furnace showed, the heat of transformation raises the temperature of wire nearly 150 deg. F., when continu-

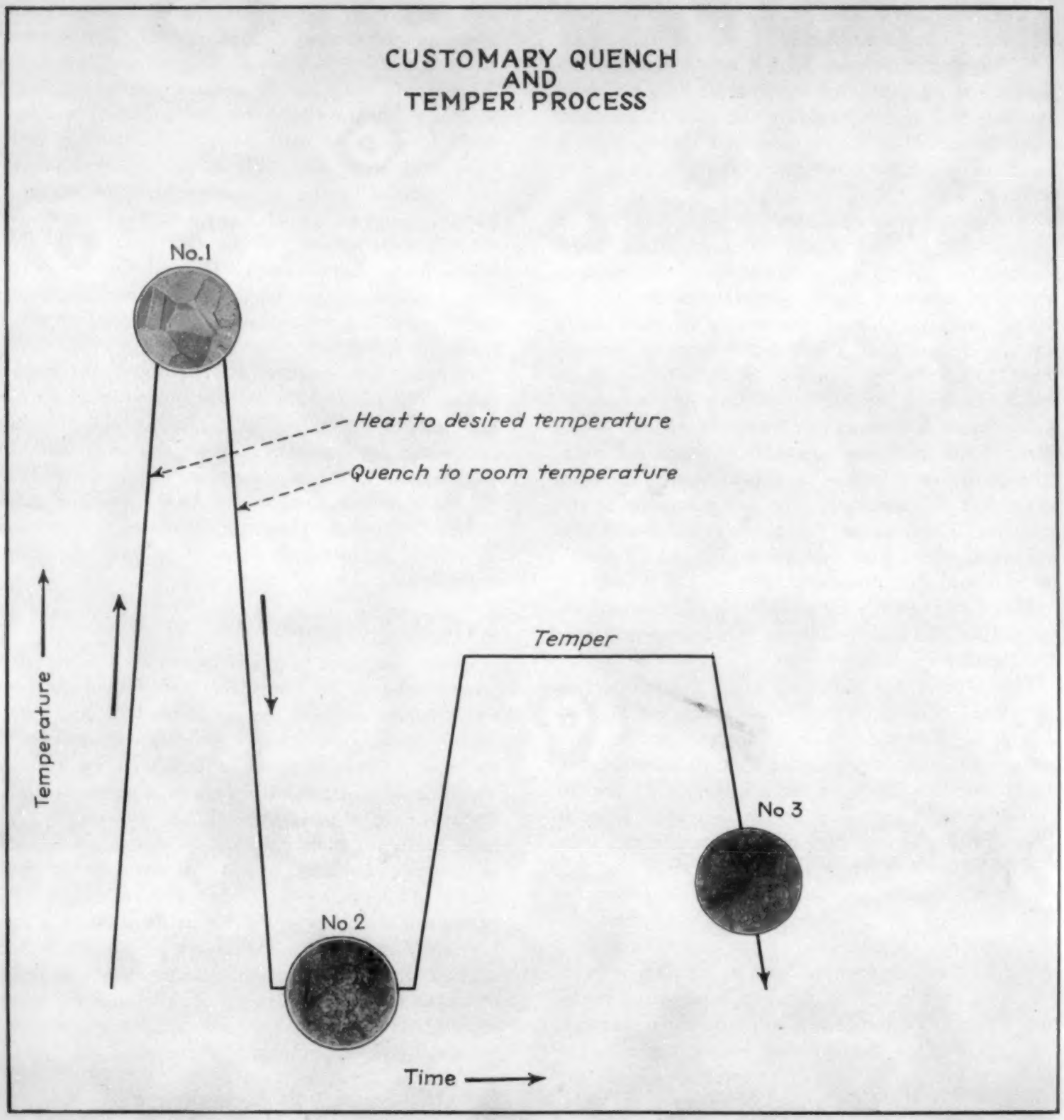


Fig. 2. Schematic Picturization of the Conventional Method of Heat Treating Steel. In this method the steel is heated to some suitable temperature above its "critical" range, until it is completely austenitic (microstructure No. 1), then quenched in water or oil, depending on its mass and composition, to give an extremely hard martensitic structure (microstructure No. 2). The steel is then tempered by reheating to some temperature between 375 and 1,000 deg. F. selected to give the structure and properties desired, before cooling to room temperature (microstructure No. 3). Control of properties of tempered steels is obtained by varying either tempering temperature or tempering time.

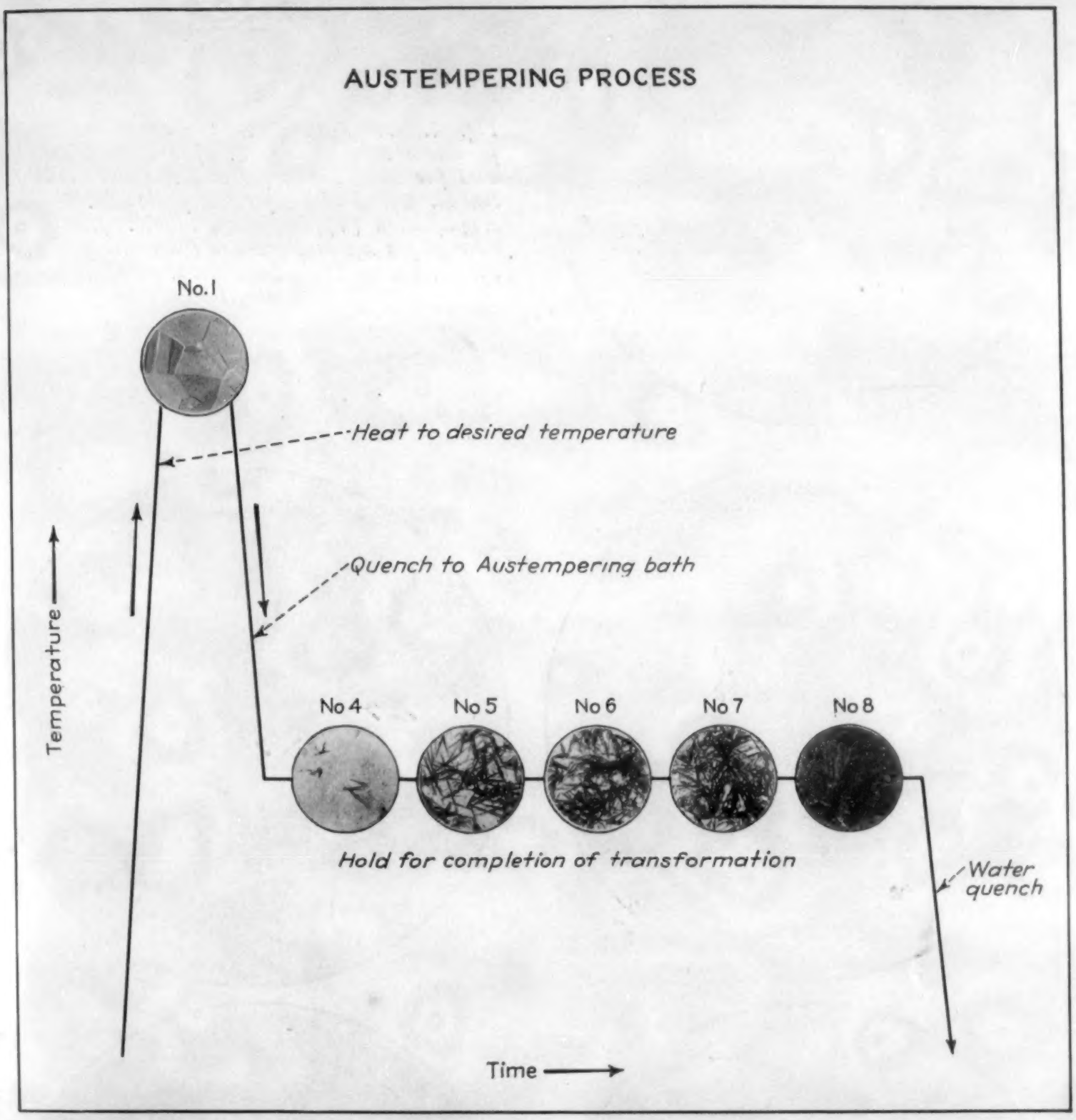


Fig. 3. Schematic Picturization of the Austempering Method of Heat Treating Steel. In the austempering method, the steel is heated until it is completely austenitic (microstructure No. 1), but instead of being quenched into a room temperature bath of oil or water, the steel is quenched into a molten salt or low-melting alloy bath held accurately at some predetermined temperature for a definite period of time known to be sufficient for complete transformation to bainite (see microstructure No. 8). The actual temperatures used range from 350 to 800 deg. F., according to the composition of the steel and the hardness desired. Microstructures Nos. 4 to 8 (reproduced in suitable size in Fig. 4) show the typical progress of the transformation of austenite to bainite at one constant temperature level.

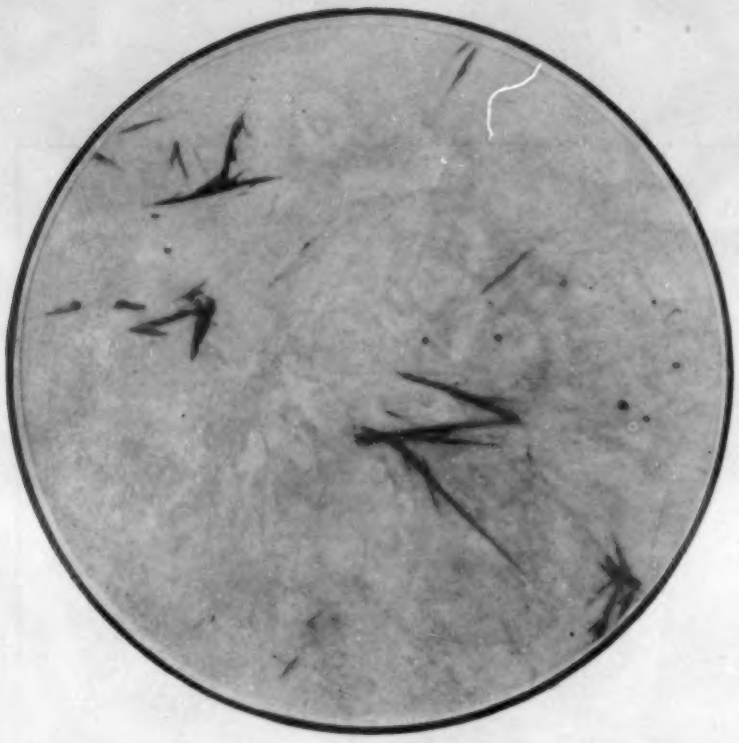
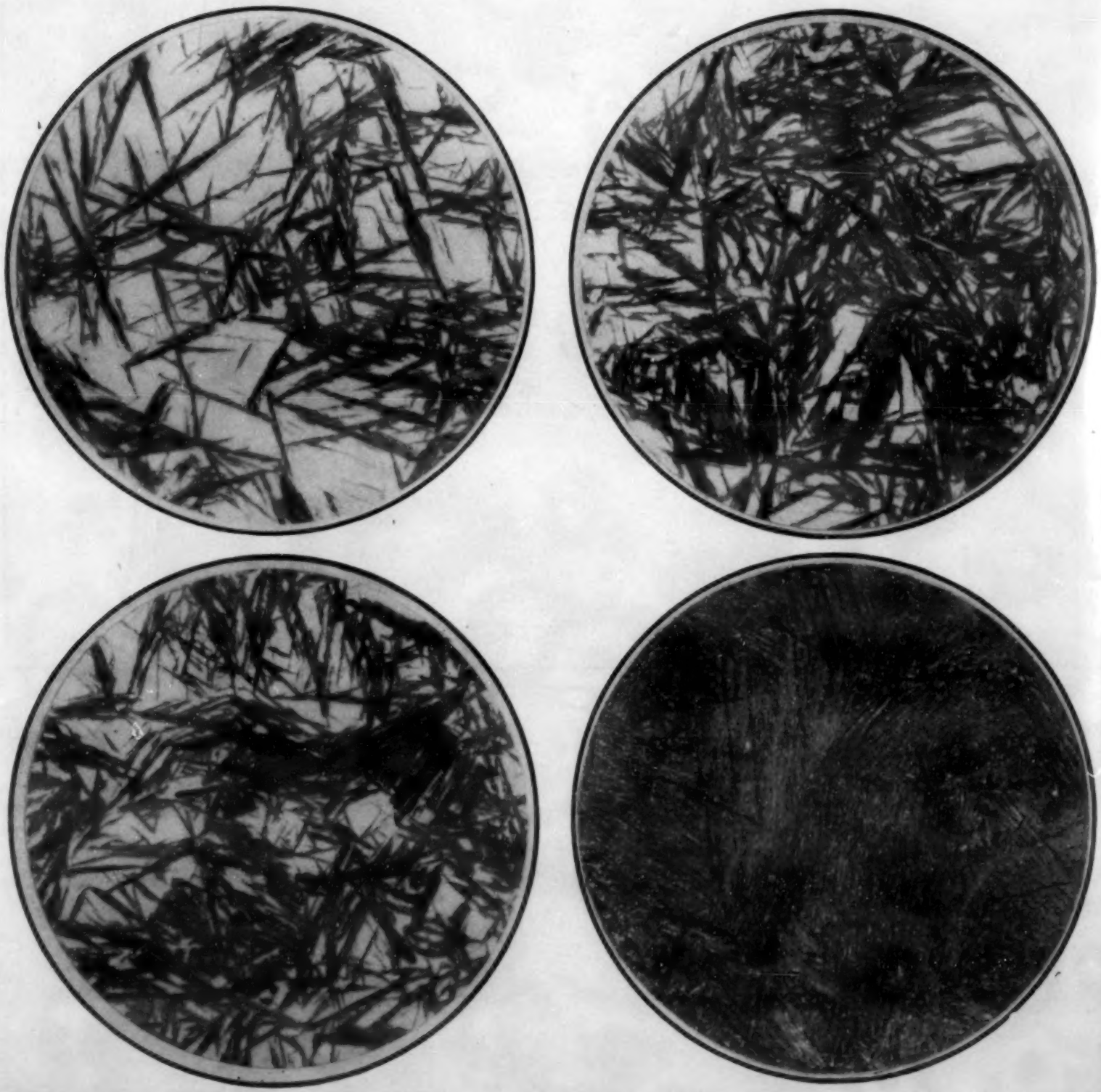


Fig. 4. Progress of Austenite→Bainite Transformation at 500 Deg. F. of a Steel Containing 0.80 per cent C, 0.64 per cent Mn and 0.166 per cent Si. Etched with 4 per cent picric acid. Original magnification (4¼-in. field), 1500X. (All photomicrographs are used through the courtesy of J. R. Vilella, United States Steel Corp. Research Laboratory.)



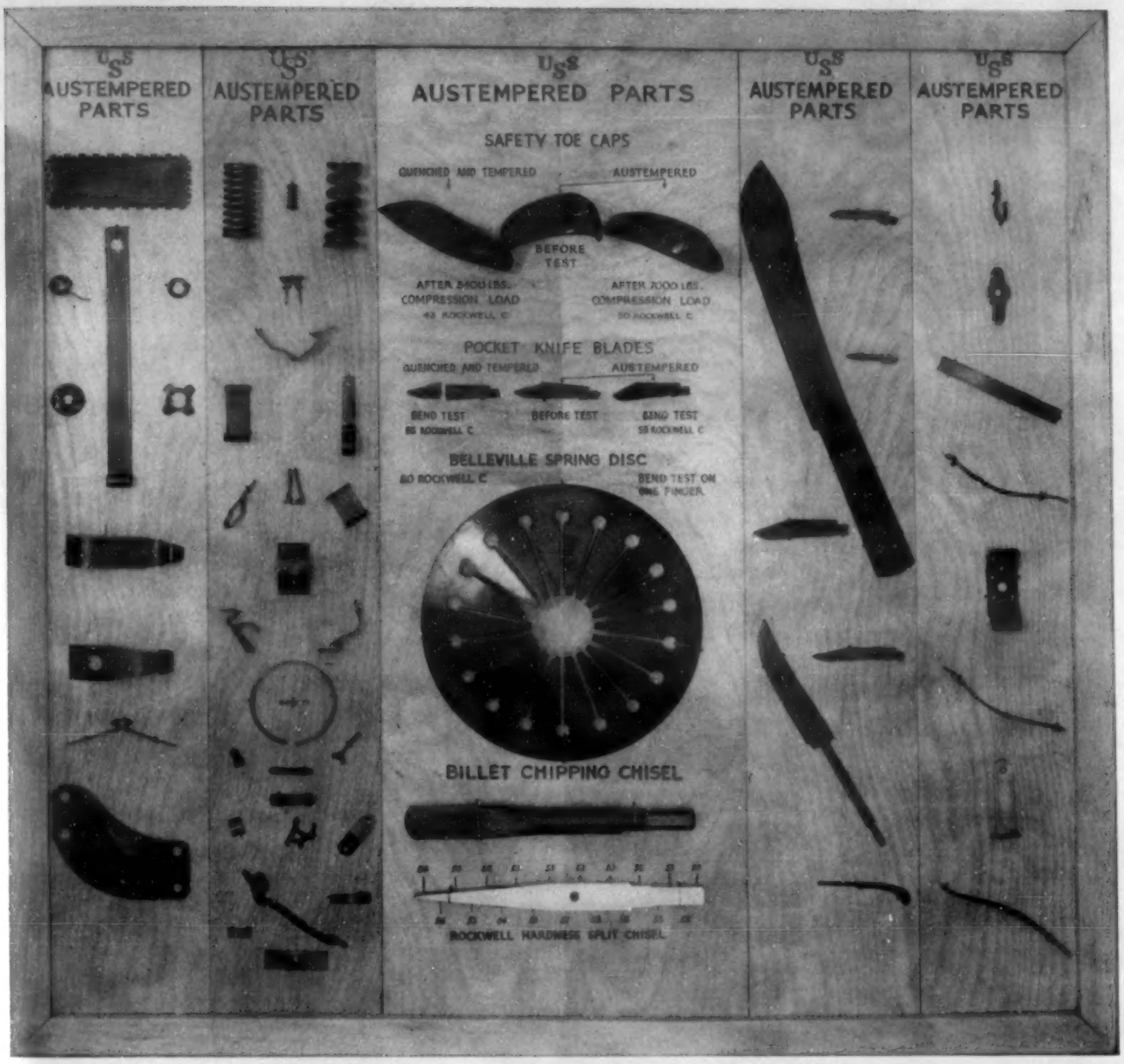


Fig. 5. Display Board Exhibited at the 1938 National Metal Exposition, Showing Some of the Products that Have Been Austempered and Indicating the Service Properties as Compared with Quenched-and-Tempered Steels.

ously austempered in air within electrically-heated tubes instead of in salt baths, and prevents uniform control of temperature. Finally, austempering of wire to the very high hardness for which the process is usually specifically employed would generally result in the rapid failure of coiling and forming tools designed to handle stock of much lower hardnesses. For these reasons, and because most of the parts that might be austempered are actually cut into individual pieces, the best initial approach was obviously to provide equipment to handle finished parts in semi-batch operation.

The furnace described and illustrated herein was built to handle a variety of problems, so that it is somewhat more elaborate than would be needed for treating a single type, or a few types, of material. Because of the longer times needed for the higher hardnesses, the salt bath must be very large regardless of the production capacity of the heating furnace. For this reason, a 500-lb. per hr. heating furnace, rather than one of 200 lb. per hr. capacity, was selected, because the investment could best be amortized with a heating furnace of high capacity. Thus, a lower cost per pound can be obtained when using moderately large hourly capacity heating furnaces, provided, of course, there is sufficient work to keep them operating somewhere near capacity.

Arrangement of Equipment

The flow sheet (Fig. 8) shows the general layout of the large scale equipment at Worcester. The work,

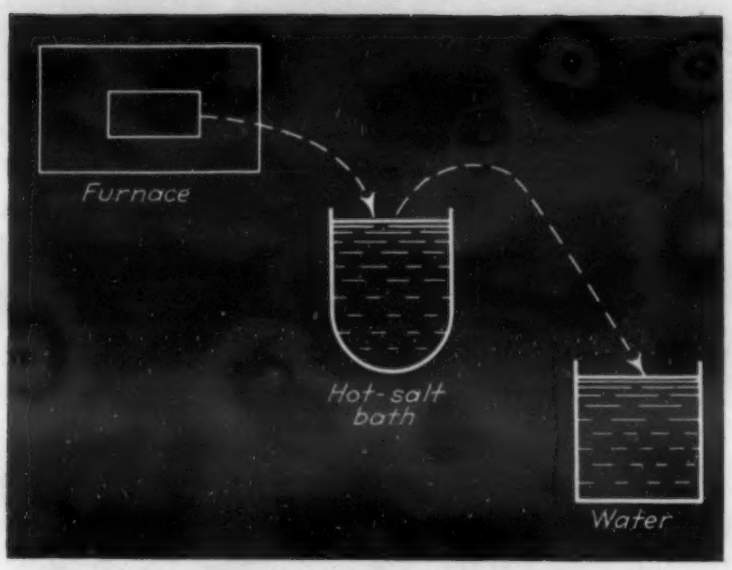


Fig. 6. Schematic Diagram of Major Equipment Units Employed in Austempering.

Fig. 7. Austempering Equipment Used for Development Work at U. S. Steel Corp. Research Laboratory.

loaded in perforated tote boxes, is placed on a gravity conveyor and is automatically elevated into the degreaser, from the upper level of which it is ejected on to a distributing conveyor before it is spilled on to the hot return belt of the hardening furnace conveyor. In passing through the hardening furnace the work is heated to a selected (austenite-forming) temperature, and is then dumped through a quenching chute on to the conveyor in the salt bath, where it is held at the desired austempering temperature. At the end of the salt bath (but still within it) the parts drop off the conveyor into suitable baskets and when the time required for complete transformation has elapsed, the baskets are mechanically lifted and dropped into a water-quenching tank. The finishing operations include washing the parts, drying by hot blast, inspection, hardness testing, and finally oiling, tumbling or ball-blasting, as the specifications require. Figs. 9, 10 and 11 are photographs of the equipment taken from various positions.

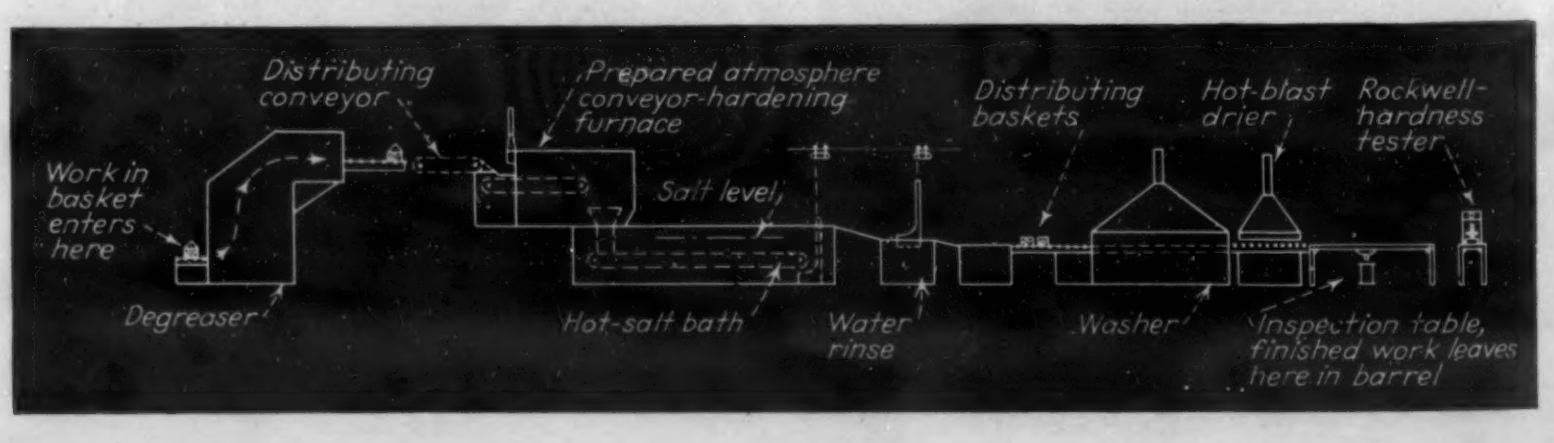
The Hardening Furnace

The hardening furnace is a gas-fired radiant tube, clean-hardening, conveyor type unit, designed for batch operation, with a capacity of 500 lbs. per hr. The operating temperature is usually held at 1575 deg. F. to secure nearly complete solution of all carbides generally present in the annealed or spheroidized steels used for cold forming. A fine-grained steel is preferred for this work, and therefore no austenitic grain coarsening results from the relatively high hardening temperature.

This temperature is controlled in four zones, all recorded: One at the entrance, one in the center, one at the discharge end as the work drops into the salt bath, and one over the salt bath chute to correct for black body radiation. The radiant tubes completely encircle the furnace conveyor belt, which is made of 25 Cr, 20 Ni alloy mesh having a total width of 24 in., of which 18 in. is actually used in operation.

The effective heating length of this belt is 9 ft., and the maximum clearing height at the entering door is 2⅞ in. The mechanical limitations on sizes of work the furnace can handle range from a round pin of 3/32 in. diameter and ½ in. long to a disc

Fig. 8. Flow Sheet of the Commercial Austempering Installation at American Steel & Wire Co.'s Worcester Works.



16 in. in diameter with a "formation" no higher than 2⅞ in. Thinner sections than 3/32 in. in flats may be handled, down to 1/64 in. thick, if general shape and mass provide enough rigidity to permit hardening in batch type equipment. Adjustment of the speed of the conveyor permits variation of heating time from 5 min. to 25 min., depending on the size and composition of the work.

The furnace shell is sand-sealed. The atmosphere employed to obtain clean hardening is prepared from a 4,000-cu. ft. per hr. unit, and refrigerated to a dew point of 34 deg. F. A direct-reading (thermal-conductivity type) atmosphere analyzer, regularly calibrated against Orsat gas analyses, is used to record the constancy of the furnace atmosphere and to indicate the need for adjustment. The average analysis of the prepared atmosphere is 9-10 per cent carbon monoxide, 7-8 per cent carbon dioxide, 8 per cent hydrogen, 1 per cent methane, remainder nitrogen. There is, of course, a very small amount of water vapor, but the CO/CO_2 and H_2/H_2O ratios in the furnace are such that the steel comes out bright, and with decarburization held at an absolute minimum.

The chute from the hardening furnace extends into the large salt bath below, making a perfect seal. The material of the chute is an alloy steel; ordinary steel would tend to catalyze the breakdown of carbon monoxide to carbon at temperatures below 1300 deg. F., and at the quenching surface of the salt bath such elemental carbon might be ignited by the hot nitrate, with disastrous results. An alloy slide is incorporated in the chute to be used when treating large pieces. Small parts are allowed to drop directly into the salt by shifting the slide into a forward position.

The Salt Bath

The salt bath is about 35 ft. long, 3.5 ft. wide and 2.5 ft. deep. It holds about 70,000 lbs. of the salt mixture, which consists of equal parts of sodium nitrite and potassium nitrate. Heating is accomplished by many gas-fired radiant tubes, which are so designed that they may also carry air and thus serve as cooling units when required. The end of the salt bath into which work from the hardening furnace is quenched is enlarged to permit rapid extraction of the extra heat introduced by the work at this point; auxiliary cooling tubes in this enlarged section carry air from a separate blower operated by a control when the temperature exceeds the "neutral" position. Four additional propeller stirrers are inserted at this location to aid in cooling.

The temperature is controlled in six zones along the salt bath as a whole, and frequent checks have shown the temperatures at the conveyor level throughout the length of the bath to be accurate within

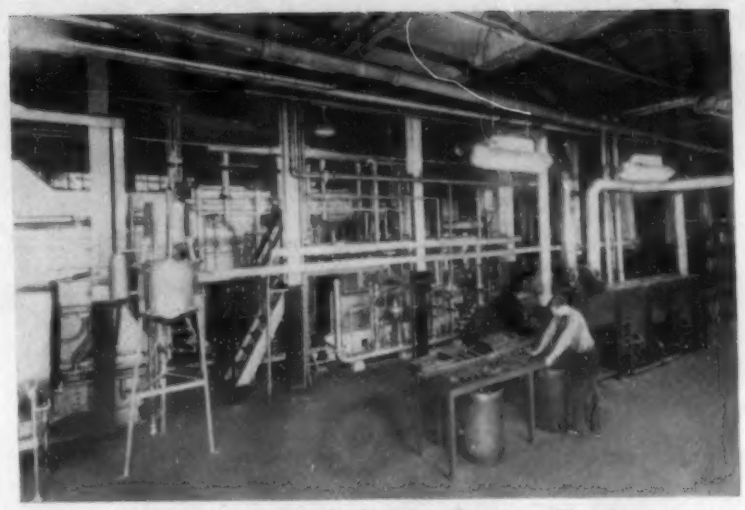


Fig. 9. A Side View of the Production Equipment at Worcester. At the extreme left are the trichlorethylene still, storage tank and degreaser. Across the rear are the hardening furnace and the salt bath; and in the right and center foreground are the washer, drier, inspection table, etc.

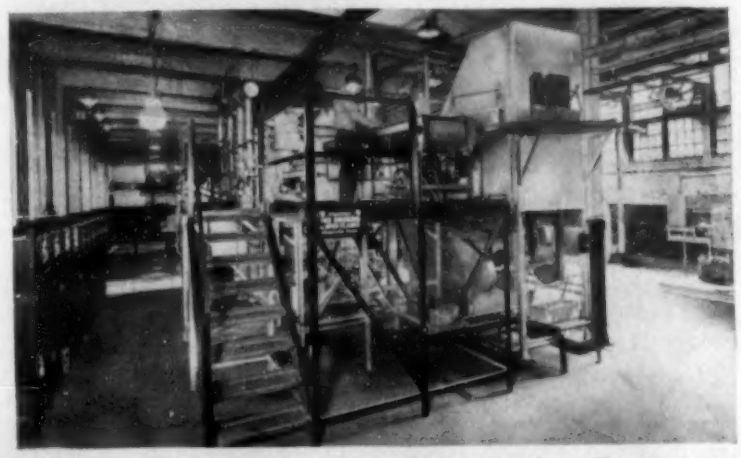


Fig. 10. The Commercial Austempering Equipment at Worcester. At the right is the degreaser, with a loading tote-box beneath it. The operator on the platform is distributing work before it enters the hardening furnace at his left. At the left can be seen the control instrument panel.

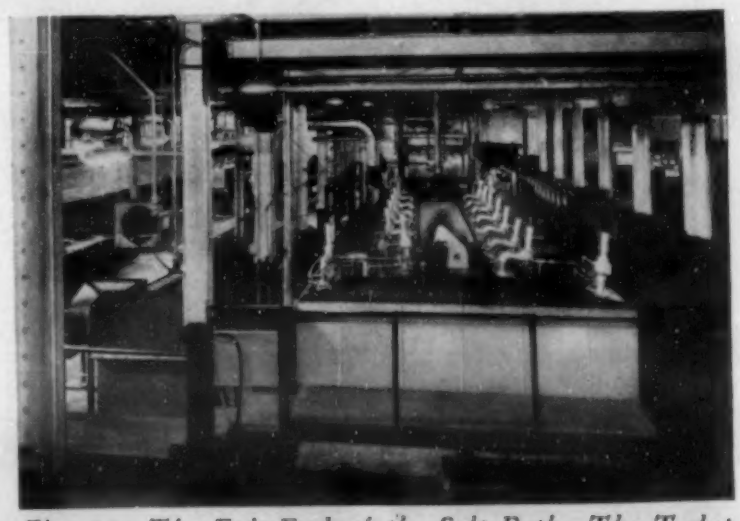


Fig. 11. The Exit End of the Salt Bath. The T-slot section of the salt bath in the foreground permits austempering for additional time after the work drops off the salt bath conveyor running down the middle of the bath. At the left is seen the water quenching tank for work leaving the salt bath.

±5 deg. F. Variable speed on the salt bath conveyor permits thin, uniform distribution of the work, and variation in time-on-conveyor from 9 to 55 min. A T-shaped section at the exit end provides means for storing the baskets into which the work falls from the salt bath conveyor for additional time in the salt, so that a total elapsed immersion time of 2 to 3 hrs. is obtainable.

For the important parts of the salt bath (except the radiant tubes, which are of 25 Cr, 12 Ni alloy) no alloy steels or irons are used, the furnace lining and conveyor being ordinary steel plate. The conveyor is operated with a ratchet arm through the top surface of the salt, eliminating the necessity for glands and bearings in the furnace itself. The intermittency of ratchet operation is so rapid that distribution of the work along the conveyor is extremely uniform. Uniform distribution is also aided by the relatively high specific gravity (2.1) of the salt bath, which permits the work to drop gently on the slowly-moving conveyor from the hardening furnace chute.

Without such even distribution, the heat of transformation of the work would cause an appreciable rise in temperature in portions of the work, a condition that must be avoided if uniformly high hardness is to be attained. Time, however, is as important as temperature in austempering and, therefore, the salt-immersion period of each batch is accurately timed. After it drops from the conveyor into baskets beneath the salt bath surface, the work may be treated for additional time, if required, merely by keeping the baskets under salt as long as is necessary, since sufficient extra space is available in the T-slot section at the exit end of the bath.

Any work carried under the conveyor and dropped on the floor of the salt bath can be removed with a sealed electromagnet. For making repairs the entire salt bath conveyor may be removed as a unit by an overhead monorail. If it is necessary at any time to dump the contents of the salt bath, a low-alloy, high strength steel basin (visible in the foreground of Fig. 11 and around the sides of the bath) has been installed that will hold the entire 70,000 lbs. of salt.

The natural finish of the austempered work after washing and drying depends on the initial finish it possessed before entering the hardening furnace. Surfaces originally reasonably bright will come out of the austempering treatment with an excellent bright blue finish that requires no further processing.

Auxiliary Equipment

No description of a heat-treating process that fails to cover the numerous auxiliaries employed can be called complete, for today's metallurgical engineer concerns himself not only with metals and the major equipment for treating them, but with all the supplementary apparatus and devices as well. Most important of these in the austempering set-up at Worcester are the degreaser, the instrumentation system, the air-moving features, and the general handling arrangement outside of the furnaces.

The degreaser is of the trichlorethylene type and is provided with a storage tank and still with automatic temperature control. This particular degreaser serves a three-fold purpose: (1) To elevate the work to the high platform on a level with the charging door of the hardening furnace, which furnace must be high enough to permit *dropping* heated parts into the salt bath; (2) to clean the work for the sake of the ultimate finish; and most important, (3) to remove surface oil that might enter the salt bath as elemental carbon and cause rapid ignition and burning. Although this is not considered a really serious hazard, the equipment has been engineered to eliminate the possibility of its occurring.

There are many other safety precautions. For example, the tote-boxes are automatically ejected from the degreaser on to a gravity conveyor, tilted to a 12-ft. composition belt conveyor with variable speed control so that work may be uniformly distributed, and spilled onto the hardening furnace hot return belt while the operator does other things. Manual loading is thus done at a safe distance from the hot belt where the furnace gases are burning prior to exhaustion. Additional precaution is afforded by a mine alarm at this point to prevent the accumulation of an unnoticed excess of carbon monoxide from either the furnace atmosphere outlet or from faulty burner connections.

Fig. 10 shows the entire control panel at the extreme left: Four recording controllers for the hardening furnace; the analyzer for the furnace atmosphere; an additional pyrometer with an alarm light set at 50 deg. F. higher than the salt bath temperature to warn of possible carbon-ignition at the surface of the salt bath; and six recording controllers for the salt bath. These can be seen more clearly in the right-hand corner of Fig. 11. The recording controllers command combination air and gas, motor-operated valves.

For supplying air to all burners on both furnaces, to the radiant tubes on the salt bath when used for cooling, and to the atmosphere-protecting eductor stream at the charging end of the hardening furnace, a blower having a maximum volume capacity of 430 cu. ft. per min. of air at a pressure of 24 oz. per sq. in. is employed. An additional blower, of 550 cu. ft. per min. capacity and 16 oz. per sq. in. pressure, supplies air to the special cooling tubes in the enlarged quenching-chute section of the salt bath. One of the interesting features of the air system is the use of the salt bath cooling air, after it has been passed through the cooling tubes and has been thereby warmed, to supply heat to the basket reheater that

prepares baskets for reinsertion in the exit end of the salt bath.

When the baskets, loaded with work, are removed from the salt bath at the end of the austempering treatment, they are quenched in a water bath (at left of Fig. 11) that removes most of the adhering salt. To keep the water-quenching bath fresh, the salty water is constantly pumped back to a salt reclaimer near the salt-quenching end of the hardening furnace (center of Fig. 11). The work is then dumped on a redistributing table from which it is conveyed automatically through a hot water spray washer, a cold water shock and rinse, a final hot water spray, a hot air blast drier, and on to the inspection table and the finishing operations mentioned earlier.

Some Operating Cost Factors

Two operators are needed on each shift to run the furnaces and auxiliaries. It has been found most economical to keep the salt bath at least molten over week-end shut-downs. The usual practice is to heat the bath to the maximum temperature possible (about 800 deg. F.) before shutting down, then permitting it to cool at about 20 deg. per hr., and later to reheat and recool it if the shutdown period is protracted. The hardening furnace is turned off completely during shutdown periods.

In normal operation, the amount of 530-B.t.u. gas consumed in austempering about 500 lbs. of steel per hr. is roughly as follows: For heating the hardening furnace, 1200 cu. ft. per hr.; for heating the salt bath, 1300 cu. ft. per hr.; for making the gas atmosphere, 1000 cu. ft. per hr.; total, 3500 cu. ft. per hr. The salt bath, of course, operates at various temperatures depending on the hardness desired, so that gas consumption for this unit may be higher or lower than the foregoing figure at different times. There is also available excess fuel capacity up to 4000 cu. ft. per hr., to permit altering the salt bath temperature at a rate of about 30 deg. per hr.

Taking all factors into consideration — original costs, operating costs and reasonable amortization— the estimated cost of austempering on a scale about as large as the one described is approximately the same as it is for the conventional quench-and-temper method.

Metallurgical Factors

The many metallurgical factors—composition, mass and austenitic grain size of the steel, and its cooling rate from the hardening (austenitic) temperature— that influence the process and result of the customary quench-and-temper treatment of a given steel also affect its treatment by austempering. The incidence of these factors, however, is different in austempering, and certain special precautions peculiar to the new process are essential to its success.

A great body of industrially useful published knowledge has accumulated on the conventional method as applied to scores of plain carbon and alloy steels, but the large amount of practical data on austempering that the heat treater will need to apply to a variety of steels and under different conditions are only now becoming available. For example, the successful application of austempering to any steel depends on a thorough knowledge of its constant-temperature transformation characteristics; in other words, part or all of the S-curve for each type of steel must be worked out, because changes in composition may considerably shift the location of the S-curve along the time axis and have a tremendous effect on the transformation time interval at certain temperatures. Some of these various S-curves have been published; several more, not yet published, have been worked out and are now available for process control; still others will be developed as time goes on.

For the present, the diagram of Fig. 12 will aid in demonstrating the general effects of composition and cooling rate on hardness obtained (incidentally, by either the conventional or the austempering method). Successful austempering of a given steel depends on a knowledge of the "nose-time" of its S-curve and of the cooling rate across the whole section of a treated part. For example, carbon steel C in the diagram, cooled from its austenitic range at a cooling rate corresponding to the quenching curve No. 4, successfully passes the S-curve nose-tempera-

Table I—Austempered Hardness Ceilings and Suggested Maximum Austemperable Sizes for Various Carbon and Alloy Steels

Type of Steel	Composition, per cent					Hardness Ceiling Rockwell "C"	Max. Dia. Round (or equivalent square), in.
	C	Mn	Cr	Ni	Mo		
Carbon	0.95–1.05	0.30–0.50	.....	.....	.....	57–60	0.148
Carbon (High Mn)	0.95–1.05	0.60–0.90	.....	.....	.....	57–60	0.187
Carbon	0.80–0.90	0.30–0.50	.....	.....	.....	55–58	0.156
Carbon (High Mn)	0.80–0.90	0.60–0.90	.....	.....	.....	55–58	0.218
Carbon	0.60–0.70	0.60–0.90	.....	.....	.....	53–56	0.187
Carbon (High Mn)	0.60–0.70	0.90–1.20	.....	.....	.....	53–56	0.281
Carbon (Very High Mn)	0.60–0.70	1.60–2.00	.....	.....	.....	53–56	0.625
Alloy	0.65–0.75	0.75–0.95	.....	.....	0.25	53–56	0.625
Alloy (C-Cr 10100 K)	1.00	0.40–0.60	0.40–0.60	.....	.....	57–60	0.312
Alloy (S.A.E. 4150)	0.45–0.55	0.60–0.90	0.80–1.10	.....	0.15–0.25	52	0.500
Alloy (S.A.E. 4365)	0.60–0.70	0.50–0.80	0.50–0.80	1.50–2.00	0.30–0.40	54	1.0 or larger

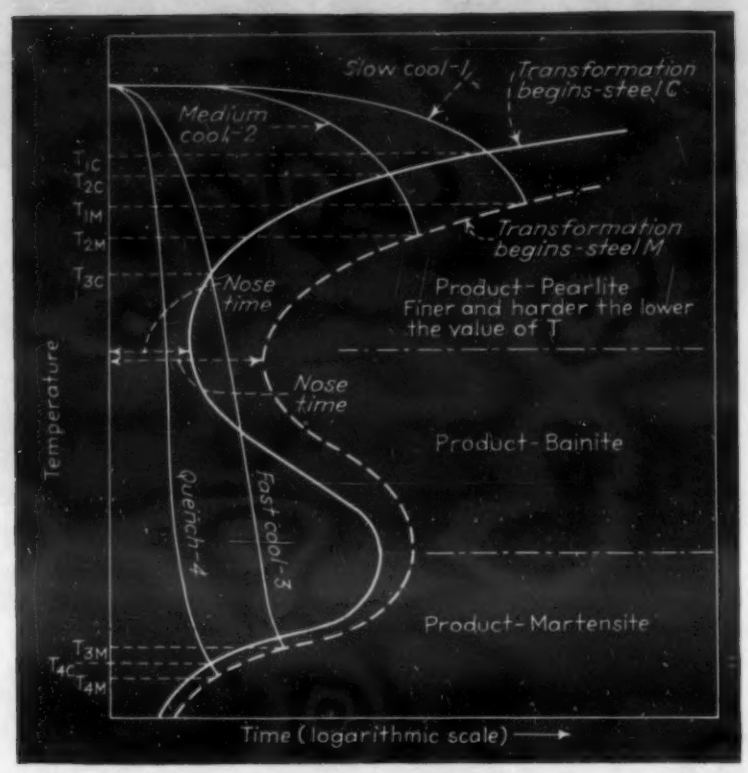


Fig. 12. S-Curves for Two Different Steels, Showing the Effects of Composition and Cooling Rates on Structures (and therefore Hardness) Obtained. These are not complete S-curves, that is, only the times required for beginning of transformation, and not the total transformation times, are shown.

ture without transforming, so that the austenite may be transformed at any desired temperature below this. Cooling or quenching rates for the same steel represented by Curves 1, 2 and 3 (which might correspond to the cooling rates near the center of a large section), however, fail to clear the nose-time of the curve for steel C, and the transformation product is pearlite instead of the desired bainite. When this happens the fundamental purpose of austempering—to obtain complete transformation of austenite *directly* to hard, tough bainite—is defeated.

The substitution, however, of steel M with a low alloy content for steel C in sizes that have a quenching rate corresponding to Curve 3, will permit such a cooling rate to clear the nose of the S-curve, for steel M's S-curve is located (along the time axis) to the right of that of steel S. In fact, quenching rate 3 for steel M clears the nose-time of the S-curve just as well as did the much faster quenching rate 4 for steel C. However, cooling rates 1 and 2 do not clear the nose-time of the S-curve of either steel C or M, and such steels, in sections so large that central portions, even under drastic external quenching conditions, could cool no faster than rates 1 and 2, would not be austemperable.

There is, therefore, a critical size for each composition of steel above which austempering is not completely successful. As the critical size for each composition is reached, attempts to austemper the steel at relatively high constant-temperature levels result in the formation of small amounts of "nose-product" (fine pearlite), although constant-temperature transformation at lower temperatures may still produce 100 per cent bainite.

Inasmuch as alloy compositions have longer nose-times, and therefore, can be austempered in heavier sections, it may be asked why such compositions are not used in all cases. Unfortunately, complete S-curves of alloy steels show that not only their nose-times, but the total time periods required for transformation are increased, in many cases, to such an extent as to affect the economy of the operation. Steels, therefore, with S-curve nose-times just safely sufficient to permit austempering a given size are most desirable; those containing small amounts of alloying elements, chiefly manganese, molybdenum, chromium and nickel, are generally selected for austempered parts of larger sizes than are feasible with plain carbon steels.

Specific Size Limits

Experiments made to date, supported by considerable practical experience on the production equipment at Worcester, permit expression of size limits and "hardness ceilings" for certain austempered steels. Table I is, therefore, offered as a guide to possible applications of the process to steels that are economically suited to austempering. The ceiling of hardness is simply the highest austempered hardness that can be obtained for each steel by constant temperature transformation at the relatively low temperature levels. Because austempering is usually applied to obtain high hardness, among other things, the application of the process to plain carbon steels of less than 0.60 per cent C is not recommended, although 0.55 per cent C steel can be successfully treated.

The effects of manganese and chromium, individually and collectively, are very strong, and therefore, the sizes given in Table I should be interpreted as applying to the mean analysis values. Variations toward the low and high side of the standard analysis specification result in some modification (at least 10%) of the maximum size austemperable.

The phrase "equivalent square" used in the table means a square section equivalent in area to that of the round. It does not follow that the maximum thickness of flat stock that can be austempered is in a simple ratio with the maximum diameter of round stock of a given steel. The larger the *volume,* the thinner are the maximum sections in a given composition that can be austempered without the formation of nose-product. For example, an 0.80 per cent C steel with 0.80 per cent Mn can be austempered in a 0.218-in. round and also in a 0.500-in. by 0.156-in. flat, but in a 10-in. disc the maximum austemperable thickness is only 0.062 in., unless portions of the center are cut away as in a belleville spring.

By prequenching work at a salt bath temperature level lower than that of transformation and quickly removing it to the desired transformation tempera-

ture, sufficient increase in cooling rate is obtained so that slightly larger sections than those given in the table can be austempered. In such cases, however, a slight admixture of low-temperature bainite is often obtained, which is subsequently tempered to a softer structure at the major transformation temperature.

Mechanical Properties

In Table II are presented the results of mechanical property tests made on the same steel heat treated in the usual way on one hand, and by austempering on the other. Table III also gives similar comparisons of mechanical properties at several different hardness levels.

The tensile strengths obtainable by austempering to a given hardness are of the same order as those resulting from heat treating in the regular way to the same hardness. In hardnesses lower than 60 Rockwell C, tests show an average of one-half a point Rockwell C higher for austempering, for equivalent tensile strength. The yield ratio for austempered material, however (not given in the table), is apparently slightly less than for quenched and tempered steel of the same hardness. Austempered steels should be treated to hardnesses one to two points higher than those of the same steel conventionally heat treated if the products are to have equal yield strengths or elastic limits.

The improvement in elongation for equal tensile strengths by austempering may be as high as 30 per cent for high hardnesses. In the low hardness ranges the elongation of austempered steels is little, if any, superior to that of quenched and tempered steel of the same tensile strength. Austempering, however, brings a very great improvement in reduction of area and, therefore, in rupture stress. Thus, Table II shows a reduction of area for austempered steel of 46.4 per cent in comparison with 26.1 per cent for the same steel quenched and tempered to the same hardness.

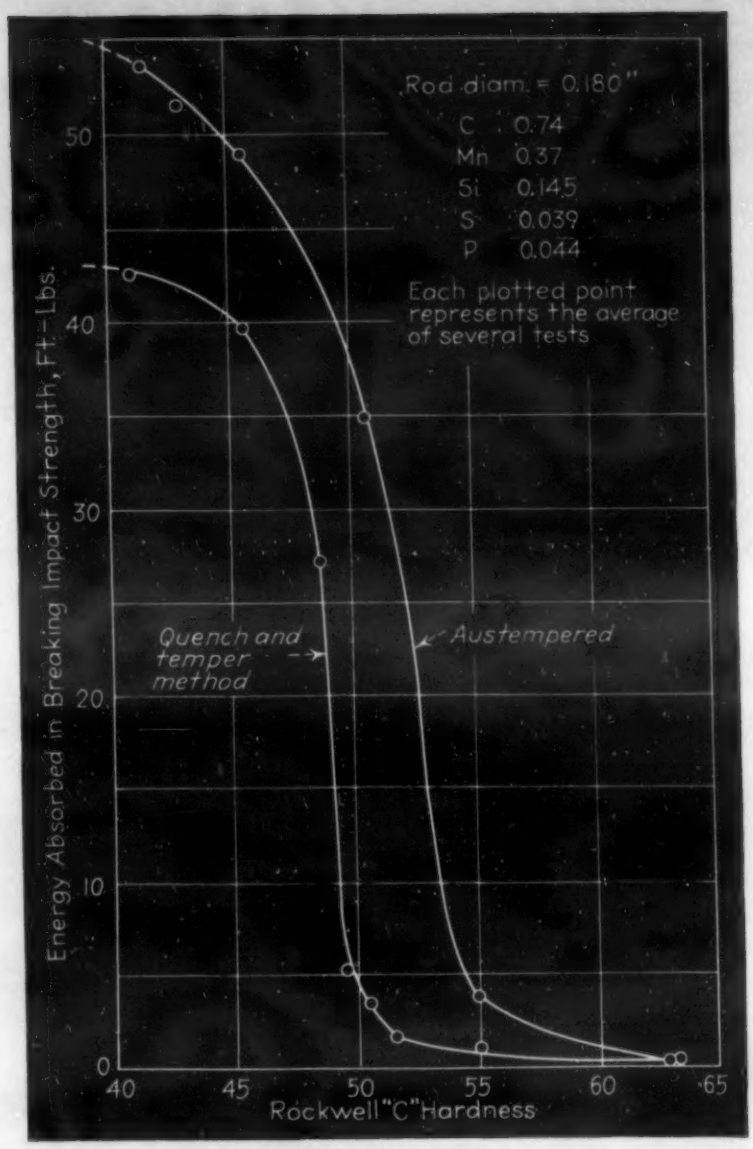


Fig. 13. Comparison of Impact Strengths at Various Hardnesses of Quenched-and-Tempered and Austempered Steels. (Courtesy United States Steel Corp. Research Laboratory)

Table II—Summary of Tensile and Impact Properties of 0.180-in. Round Rods Heat Treated by the Quench-and-Temper Method and by Austempering. All results except those of the bending tests are the average of three determinations. (Courtesy United States Steel Corp. Research Laboratory)

Property Measured	Quench-and-Temper Method	Austempering
Rockwell C Hardness (flat surface)	49.8	50.1
Brinell Hardness (converted).....	486	489
Ult. Ten. Str. lbs./in.² (computed on original cross-section)	259,000	259,300
Rupture Stress, lbs./in.² (computed on reduced cross-section).....	312,150	355,500
Elongation, per cent in 2 in.....	3.75	5.0
Red. Area, at fracture, per cent....	26.1	46.4
Impact, ft.-lbs. (90 ft.-lb. hammer) Unnotched round specimen.....	14.0	36.6
Bending (free bend in ¾-in. length)	Ruptured at 45°	Greater than 150° without rupture

Analysis: C, 0.78; Mn, 0.58; Si, 0.146; P, 0.042; S, 0.040.

Heat Treatment:

Quenched-and-Tempered	Austempered
Pb bath 1450° F.—5 min.	Pb bath 1450° F.—5 min.
Oil quench	Transformed in Pb-Bi bath
Tempered 650° F.—30 min.	at 600° F.—20 min.

Grain Size (1450° F.): 5-6, with 6 predominating.

Impact strength, tested on unnotched round specimens, also shows a striking improvement by austempering, particularly for steels treated within that hardness range (Rockwell C 48-55) for which austempering is most advantageous. This is demonstrated not only in Table II but in Table III and Fig. 13 as well. Table III and Fig. 13 also show impact strength superiority for austempered steel in hardnesses below 48 Rockwell C. This is small, however, as would be expected in view of the improved ductility of quenched and tempered steel in these lower hardnesses.

Slow bend tests also indicate the extreme toughness imparted by austempering, particularly in hardnesses between 48 and 60 Rockwell C. Fig. 14 shows comparative slow bend test results obtained at two hardness levels.

Interpretation of fatigue results must be done cautiously, in view of the many factors other than the composition and heat treatment of the steel which may affect fatigue behavior. The results of endurance

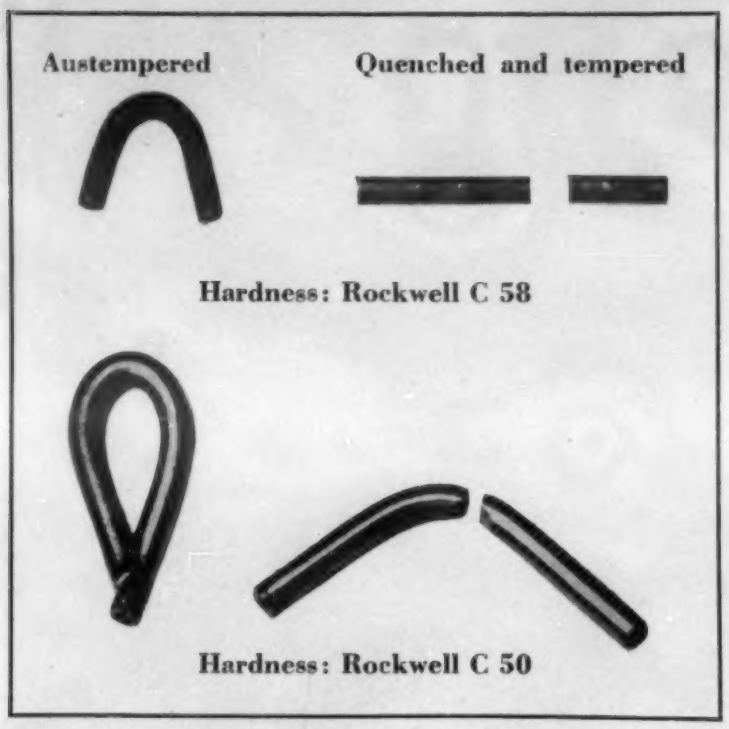


Fig. 14. Results of Slow Bend Tests on Quenched-and-Tempered and on Austempered Steels. (Courtesy United States Steel Corp. Research Laboratory)

Table III—Summary of Mechanical Properties of 0.180-in. Dia. Carbon Steel Rods, Heat Treated by Quench-and-Temper Method and by Austempering. (Courtesy United States Steel Corp. Research Laboratory)

Property Measured	Quench-and-Temper Method	Austempering	Analysis, per cent C	Mn	Si
Hardness (Rockwell C, flat surface)	50	50	0.85	0.42	0.180
Yield Str., lbs./in.², 0.1% set	224,750	210,950			
Tensile Str., lbs./in.²	261,100	256,700			
Elong. in 2 in., per cent	4.7	6.2			
Red. Area, per cent	28.3	44.8			
Impact (ft.-lb.)*	20.4	40.2			
Hardness (Rockwell C, flat surface)	45	45	0.74	0.37	0.145
Tensile Str., lbs./in.²	250,000	254,300			
Elong. in 2 in., per cent	6.0	6.3			
Red. Area, per cent	42.5	49.0			
Impact (ft.-lb.)*	41.1	No fracture			
Hardness (Rockwell C, flat surface)	41	41	0.74	0.37	0.145
Tensile Str., lbs./in.²	226,300	224,700			
Elong. in 2 in., per cent	8.8	9.0			
Red. Area, per cent	42.7	54.2			
Impact (ft.-lb.)*	34.5	50.1			
Hardness (Rockwell C, flat surface)	46	46	0.62	0.80	0.171
Tensile Str., lbs./in.²	253,700	248,000			
Elong. in 2 in., per cent	7.0	7.3			
Red. Area, per cent	46.2	50.4			
Impact (ft.-lb.)*	35.9	39.6			
Hardness (Rockwell C, flat surface)	41	41	0.62	0.80	0.171
Tensile Str., lbs./in.²	223,700	225,000			
Elong. in 2 in., per cent	8.0	7.5			
Red. Area, per cent	50.3	56.4			
Impact (ft.-lb.)*	37.4	45.9			

* Unnotched specimens, 0.180 in. diameter.

tests made on a rotating beam machine, given in Fig. 15, show almost identical values for austempered and quenched-and-tempered steels of equal hardnesses; this is not surprising inasmuch as all samples were machined to the same "perfect" surface and had been treated to the same hardness. Improved fatigue properties should not be expected in austempered steels just because of the higher hardness-toughness combinations obtainable with them. In actual service, surface defects of a stress-raising nature may react more strongly on the austempered steel only if it is harder, because of the generally greater notch sensitivity of higher hardness materials. In the same way, the conventional practice of improving endurance by shot-blasting to put compressive forces on the highly stressed surface of parts used in torsion or bending can be effective only on austempered parts softer than 50 Rockwell C. Steels harder than this evidently resist the surface deformation normally induced by ball-blasting to such an extent that improvement in fatigue properties is negligible. Finally, decarburization during austempering must be avoided, if for no other reason than its bad effect on fatigue behavior.

Fatigue tests run on production material with natural finishes have shown superiority for quenched-and-tempered steels in some cases and for austempered in others. This variation in results is quite likely because other factors in such tests have more to do with the fatigue behavior of the material than does the method by which it was heat treated. Fatigue results on austempered steels, however, usually deviate less from their average than do those on quenched-and-tempered; this is probably a reflection of the greater uniformity of hardness obtained by austempering.

Practical Applications

A few practical illustrations of the service to which austempered steels can be subjected may indicate the possibilities in the process. Thus, safety toe caps for factory and warehouse workers' shoes can be austempered to a higher energy-absorbing value than if quenched and tempered. Fig. 16 demonstrates this graphically. In this case the use of the process means caps that support a greater maximum load and are able to absorb more energy up to the limiting dimension of a man's toe—9/16 in.

Fig. 17 shows a flexure test on a carbon steel shovel austempered to 55 Rockwell C. This is 5 to 10 points harder than can be used on shovels heat-treated conventionally, so that the austempered product offers greater wear resistance plus the reserve toughness and flexibility and freedom from distortion necessary in this type of service.

One manufacturer of large pipe wrench chains employs in them carbon-molybdenum steel pins that must be hardened to 40 Rockwell C before upsetting.

A chain, incorporating pins heat treated conventionally, was found to break at a total load of 22,000 lbs. When the pins are austempered to 50 Rockwell C, the same upsetting is possible without fracture, and the chain breaks at 28,000 lbs. total load.

Many other instances of successful application might be cited. The following list will indicate some of the products, the service properties of which are definitely improved by austempering:

Agricultural tools and implements
Automotive and aircraft parts and accessories
Bayonets, swords, etc.
Belleville springs
Business machine parts
Cotter pins
Cutlery, pocket knives, back springs
Door checks, window lifters
Electrical apparatus parts
Firearm parts
Fishing rods
Golf shafts
Hardware, miscellaneous
Industrial instruments
Last springs and shoe shanks
Letter openers
Link chains, rolls, sidebars
Lockwashers
Needles
Needle bearings
Pens
Retaining rings
Shovels and forks
Skate blades
Springs—round and flat
Textile machinery parts—spindles, bobbin rings and catches
Toe caps
Tools (small) of less than 59 Rockwell C Hardness

Many of these parts have been austempered with excellent results using plain carbon steels of 0.50 to 1.10 per cent C. When such parts are of too large section to be austemperable in carbon or carbon-manganese steels they may, of course, be successfully austempered if made of alloy steels containing small amounts of chromium, nickel or molybdenum, or further manganese additions. Certain other special alloy steel products, such as chipping chisels and pins and bolts for extra-duty service, have also been austempered to advantage.

General Appraisal

Austempering as a commercial heat-treating method has many advantages, and certain very definite limitations. Among the latter should be cited the existence of a maximum austemperable size for each steel composition, and the necessity, therefore, to modify the composition by addition of manganese or other alloying agents if larger sizes than the maximum austemperable in the usual carbon steel analysis are to be treated. Also, full advantage of the transformation-retarding propensity of more highly alloyed steels cannot always be taken because the total transformation time of such steels may be uneconomically long.

Furthermore, there are definite hardness ranges in which austempered products are considerably superior to those quenched and tempered to the same hardnesses and other hardness ranges in which the new method offers little, if any, advantage. The better properties of austempered steels are manifest in the higher hardness ranges—from 48 to 58 Rockwell C—where combinations of high hardness and toughness are eagerly sought by industry's engineers. Of particularly great value are the hardness-toughness

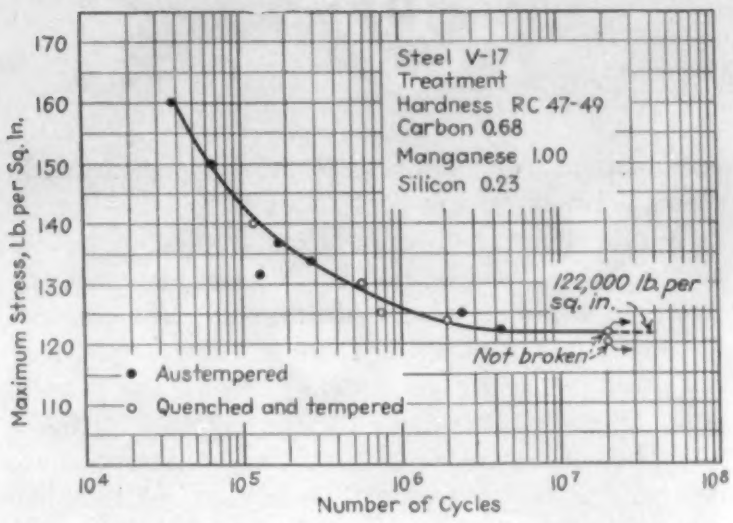


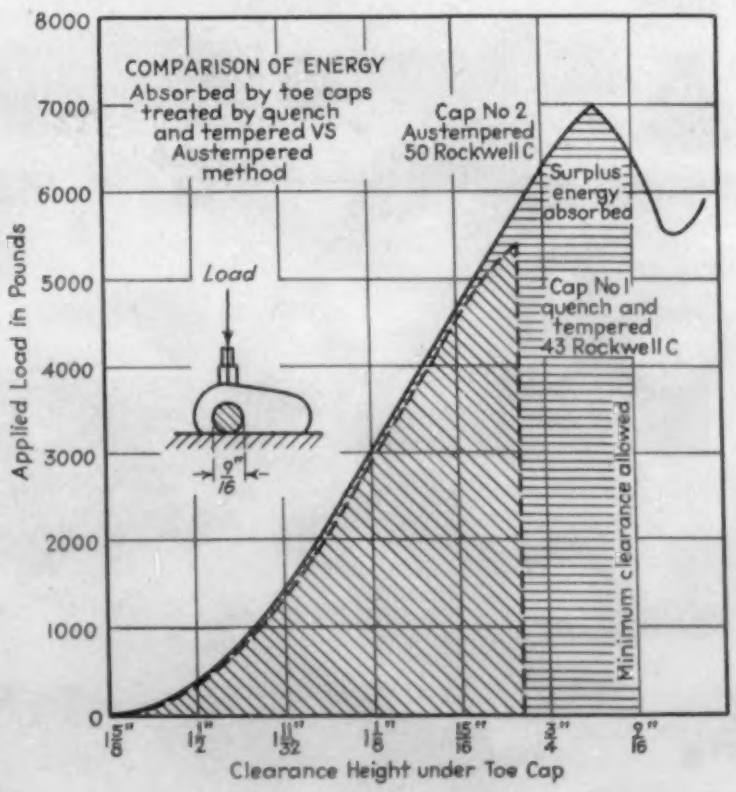
Fig. 15. Fatigue Properties of a Quenched-and-Tempered Steel and of the Same Steel Austempered. The steel contained 0.68 per cent C, 1.00 per cent Mn, 0.017 per cent P, 0.030 per cent S, and 0.23 per cent Si; all samples had identical processing history, except the final heat treatment, and were ground and polished to testing size according to the method recently described by Henderson and Seens.[5]

The final heat treatment and the resulting hardnesses and microstructures were as follows:

	Quenched-and-Tempered	Austempered
Heat Treatment..	(a) Heated 20 min. at 1450° F. (b) Quenched in oil (c) Heated very slowly to 665° F., tempered there for 45 min. (d) Air-cooled from 665° F.	(a) Heated 20 min. at 1450° F. (b) Prequenched, 10 sec. at 375° F. in Pb-Bi bath (c) Transformed at 600° F., 40 min., in salt (d) Air-cooled from salt bath
Hardness (Rockwell C).....	48.7	48.6
Austenite Grain Size	6-7	6-7
Structure	Tempered Martensite	100% Bainite

(Courtesy United States Steel Corp. Research Laboratory)

Fig. 16. Energy Absorption Tests on Safety Shoe Caps Made of Quenched-and-Tempered and of Austempered Steel. Note: Experience shows hardness above that used for quenched and tempered unsafe due to brittleness.



combinations available in steels austempered to hardnesses between 50 and 55 Rockwell C.

Within these higher hardness ranges, however, austempering offers several decided advantages in addition to the unusual combination of hardness and toughness just mentioned. There is, for example, less distortion in the process probably because the quench is not so drastic as that of the conventional method; also, there is no danger of quenching cracks occurring because the transformation from gamma iron to alpha iron of larger specific volume is accomplished *gradually*. A corollary advantage that should not be ignored is the possibility of obtaining higher hardnesses without any danger of developing quenching cracks. Steels austempered to the higher hardnesses also possess strikingly superior ductility when compared with quenched-and-tempered steels treated to approximately equal yield strengths or elastic limits, so that higher unit elasticity and improved wear resistance of a product can be attained without any sacrifice of toughness. The impact strengths of austempered steels are also much higher than those of steels conventionally heat treated to the same higher hardness levels; this opens up great possibilities for the new process in treating products that must withstand considerable wear plus occasional pounding. Finally, all this can be accomplished with, if anything, an improvement in surface finish over that resulting from the usual air-tempering operation, for there is an absolute minimum of oxidation of parts during austempering.

Finally, if the proper size of hardening furnace can be utilized to distribute the somewhat heavier investment in the salt bath than is normally encountered in oil or water baths, the overall cost of austempering should not be higher than that of an equivalent conventional heat-treating process of the same order of refinement. Considering the greater piece-to-piece uniformity and freedom from rejections caused by non-uniform hardening and distortion of individual pieces, it is felt that the austempering method has more than demonstrated its economic position.

The natural obstacle to wide adoption of the process at the present time is the relatively great additional expense involved in rebuilding or remodeling equipment that is currently operating satisfactorily. When new equipment can be considered for tonnages large enough to warrant a reasonably sized unit, the installation of this most modern of heat-treating processes can be easily justified. From the most encouraging results after 1½ yrs. of practical operation, the austempering process seems destined gradually to supersede the conventional quench-and-temper method within the metallurgical limitations and for the type of work for which it is adapted.

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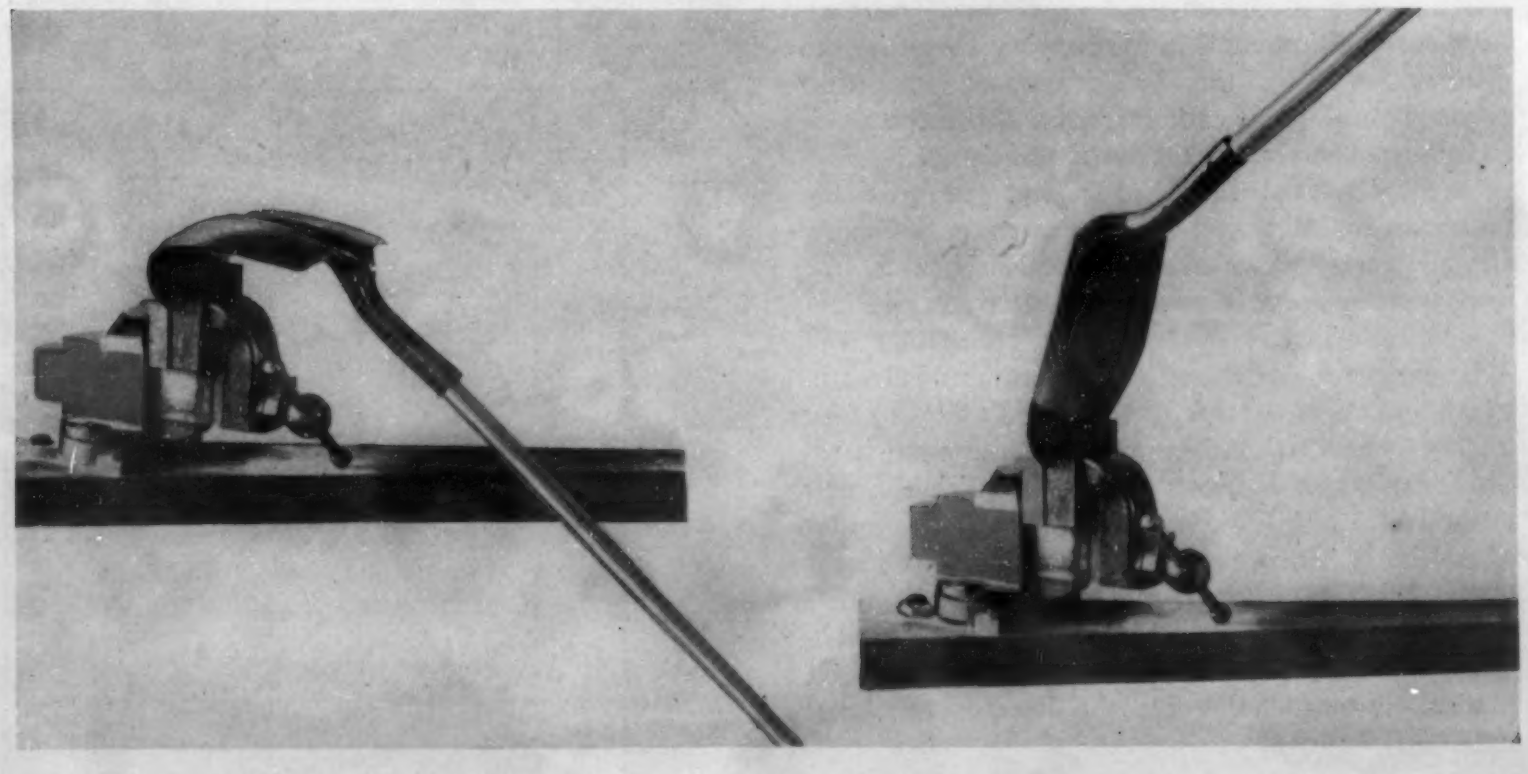
Fig. 17. Demonstration of the Resiliency of an Austempered shovel. The shovel was austempered to 55 Rockwell C, which is very high for this application. The picture on the left shows the extent to which this shovel can be bent without failure, and that on the right shows how the bent shovel, after removal of the bending force, returns to its original position without permanent deformation.



Fatigue of Light Metal Alloys

by R. L. Templin

Chief Engineer of Tests,
Aluminum Co. of America,
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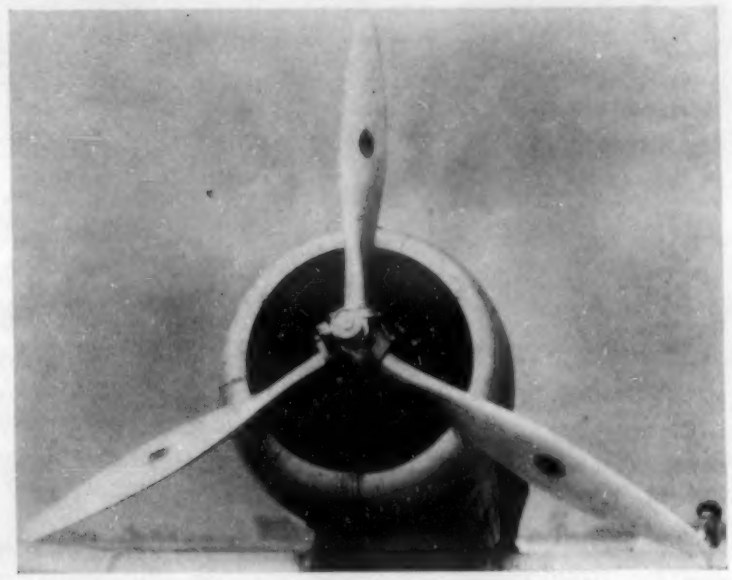

Fig. 1. Aircraft Engines and Propellers Require Special Attention in Design to Avoid Fatigue Failures.

This article is the third in a series of six on Fatigue of Metals, the first of which was published in our issues of May and June. The second on "Fatigue Problems in the Aircraft Industry" was published in the July issue.

The origin and general scope of the series was fully outlined in an introductory statement to the first installment of the first article—"Fatigue of Metals—Developments in the United States"—published in the May issue, pages 158 to 162.—The Editors.

THE LIGHT METAL ALLOYS are used extensively for many parts of machines and structures, where formerly heavier metals were used. The satisfactory results obtained under service conditions, the development of newer and better alloys, and a better understanding of the engineering design requirements involved have all helped insure a more extensive use of the light metal alloys in engineering structures.

When designing engineering structures to be built of light metal alloys, it is only natural to compare the mechanical properties of these newer metals with similar property values of the older, heavier metals. In making such comparisons certain differences will be observed. If the structures made from the newer metals are to be successful, proper allowance must be made in their design for the differences noted. The differences in the so-called "static" mechanical properties, such as occur in the case of elastic moduli, usually involve no serious difficulties in the engineering design problems presented. The differences in the "dynamic" mechanical properties, however, may cause more concern in engineering structures that are expected to withstand frequently repeated loads of appreciable magnitude, over protracted periods of time.

Experience has shown that, under such conditions, structures made from presumably well-known and widely-used materials occasionally fail to give satisfactory service. Undoubtedly the cause of such failures can often be attributed to lack of understanding of the dynamic properties of the materials and deficiency in knowledge of how to design engineering structures to withstand such types of service loads. This is frequently true even when fatigue data are available.

Alloys of Aluminum and Magnesium

In the present discussion we are primarily concerned with the fatigue properties of light metal alloys; more specifically with the aluminum-base and magnesium-base alloys. Values for some of the fatigue properties for most of the commercial alloys of these metals may be found by reference to the technical literature.[1] All the endurance limit values based on 500,000,000 cycles of stress are lower than those found for the ferrous metals based on 10 to 20 million (and probably more) cycles of stress. The life basis of 500,000,000 cycles of stress, chosen for use in reporting the fatigue properties of the light metal alloys, is simply an arbitrary one so chosen as to be directly comparable in certain cases to the probable life of the structure under consideration.

It must be emphasized that many structures made from the light metal alloys are not expected to withstand anywhere near this number of repetitions of stress. In such cases consideration should be given to the indicated fatigue strength values corresponding to the number of repetitions of stress involved. De-

pending on the number of repetitions of stress involved, quite different choices of alloys might be made, because the fatigue curves for the light metal alloys sometimes cross each other at some intermediate number of cycles; that is, for a lower number of cycles one alloy may be superior to the other in endurance strength but at a higher number of cycles the reverse may be true.[1]

After considering the points just mentioned, the question is often asked: "Cannot something be done to increase the fatigue strength of the light metal alloys?"

Factors Affecting Fatigue Values

In attempting to answer this question, very extensive laboratory fatigue tests have been carried out over the past 15 or more years, covering not only the commercial alloys but also many experimental alloys of these metals. These tests have included many factors which might affect the fatigue strengths of the alloys. Among these factors may be mentioned chemical composition, amount of cold working, various heat treatments, different fabrication processes, temperature, effects of different types of coatings, effects of various corrosive conditions, effects of various types of stress raisers, and the effects of various methods of testing on the values obtained.

In addition to the determination of the limiting values when the stresses are completely reversed, a number of laboratory tests have been carried out to determine the safe stress ranges for various mean stress values, for many of the commercial alloys.[2] Likewise, considerable attention has been given to the determination of shear fatigue values; but, unfortunately, very few values are as yet available in the field of combined fatigue stresses, for the light metal alloys.

Results obtained from these laboratory tests of the light metal alloys have formed a basis for some rather important general conclusions:

(1) The chemical composition of an alloy does have an effect on the fatigue strength of the metal, but in some instances the composition which gives the highest fatigue properties introduces other undesirable properties such as lowered resistance to corrosion or fabrication difficulties.

(2) An increase in the amount of cold working done on a light metal alloy increases the fatigue strength of the material, but not in proportion to the change in the static mechanical properties of the metal; and also increases the notch sensitivity of the material under repeated stress conditions.

(3) Solution heat treatment of the heat-treatable light alloys usually has a pronounced beneficial effect; but precipitation heat treatment appears to have but little, if any, beneficial effect on their fatigue strengths.

(4) Differences in fabrication methods used in producing the various commercial forms of a given light metal alloy produce some differences in the fatigue strength values, and much has been done in recent years to adopt processes or make changes in fabrication methods which improve the fatigue properties of the products.

(5) The serious effects of corrosion on the fatigue properties of metals are generally recognized. Consequently, much work has been done in the improvement of the inherent resistance of the light metal alloys to corrosion as well as in the development of better coatings for preventing corrosion, with resulting improvement in fatigue strength under some service conditions.

The factors just mentioned are those which are primarily the responsibility of the producers of light metal alloys and are under their control. For obvious reasons you may rest assured that they are being given all possible consideration, to the end that better fatigue resisting materials can and will be produced.

Fatigue results obtained from investigations purporting to show the effects of temperature, stress raisers and mean stresses on the safe stress ranges for the light metal alloys, although often furnished by the producers of the metals, are primarily for the use of the designing engineer. He, in the final analysis, must accept responsibility for the way in which they are used in the design of engineering structures.

Effects of Stress Raisers

Many tests have been carried out by various laboratories for the purpose of evaluating the effects of various kinds of stress raisers on the fatigue strength of metals. Those reported by R. R. Moore[3], Otto Graf[4] and Peterson and Wahl[5] are noteworthy examples. Included in the category of stress raisers are surface finish, shape and dimension of notches caused by threads, key ways and openings, as well as changes in section.

Selection of mean stress and safe stress range values, and the number of cycles anticipated in the probable life of the structure, are indeed very important. If the stresses are not completely reversed, much higher mean stress values can generally be used with safety. In the same way the use of narrow stress ranges permits higher safe mean stress values.

When a structure is expected to withstand repeated loads under service conditions, much attention should be given by the designer to the actual distribution of stress occurring in the various parts of the structure. In by far the majority of cases the actual distribution of stress will be appreciably different from the distribution assumed in the engineering design of the structure. An outstanding example of this statement occurs in the case of welded connections in structures. In such instances, the points of highly localized stresses probably have little if any effect on the static strength of the structure, but are of primary significance in the life of the structure when subjected to repeated loads.[6]

The light metal alloys have in the past found, and will continue in the future to find extensive use in certain structural engineering fields, because of their superiority in many respects to other available materials. Where their resistance to repeated loads is involved, however, and because of their relatively lower inherent fatigue strengths, suitable consideration must

be given to the design details involved in engineering structures for such service if fatigue failures under service conditions are to be prevented.

Use of Fatigue Values

Fatigue values obtained from laboratory tests of light metal alloys, like similar information for other metals, may be used (1) as a basis of comparison of materials, (2) as criteria of maximum fatigue strengths under favorable conditions, (3) as a basis for the selection of engineering design values, (4) to emphasize the necessity for avoidance of severe stress raisers in design, and (5) to show the necessity of avoiding conditions involving oft-repeated large stress ranges under service loads.

An example of what can be accomplished, in the avoidance of stress raisers by engineering design, may be found in a specific case where it was desired to substitute a forged aluminum alloy locomotive side rod for a steel side rod. Changes in design based on the usual considerations did not prevent an early fatigue failure in the cross-head end of one of the large rods. Analysis of the stresses occurring in this end of the rod led to changes in design which appreciably decreased the maximum stresses occurring; with the result that a rod was produced which, under service conditions, gave results quite comparable to those previously obtained for steel rods.[7]

As an example of what can be accomplished, in some instances, in decreasing the magnitude of stress ranges occurring under service conditions, there may be cited a problem presented by the fatigue failures of electrical transmission lines in certain localities where they are subject to unfavorable wind conditions. While selection of other materials or changes in span length, stringing tensions or design of conductor might have solved the problem, yet the most efficient and simplest solution was found to be the prevention of the excessive vibrations which occurred. This was accomplished by means of a relatively simple and inexpensive vibration damper, which through years of service has proved to be highly efficient.[8]

Vibration and Fatigue

At this point it should be emphasized that the mere presence of vibration, in an engineering structure, does not warrant the conclusion that the structure will fail by fatigue action. Only when the vibrations are of sufficient amplitude to cause stresses in excess of the fatigue strength of the material should there be cause for alarm as to the fatigue strength of the structure. In the case of the electrical conductors just cited, vibrations of very small amplitudes still persist after the application of the vibration dampers, but the stresses resulting from these minute vibrations are of no practical consequence. Vibrations undoubtedly occur in all metal aircraft, yet fatigue failures of the

Fig. 2. A Battery of Rotating-Beam Fatigue Machines Used in Studying the Fatigue Properties of Light-Metal Alloys.

essential structural parts of such structures are conspicuous by their infrequency or absence.[9]

Conclusions

In conclusion it should be emphasized that the fatigue strengths of the light metal alloys are relatively less than the values for the heavier metals. While the light alloys have been improved and are still being improved, it seems very likely that we will always be confronted with engineering problems in the design of light metal alloy structures, subjected to repeated loads, in which consideration must be given to such factors as: (1) actual maximum stresses, (2) actual stress distribution, (3) probable number of repetitions of stress expected during the life of the structure; as well as (4) the fatigue strengths of the metals. With proper attention to these factors, successful solutions of many such fatigue problems have been realized.

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A CRITICAL REVIEW • PART 4

Grain Size of Steel

(Concluded from July)

by N. F. WARD and J. E. DORN

Associate Professor of Mechanical Engineering and Assistant Professor of Mechanical Engineering, Respectively, University of California, Berkeley, Cal. Prof. Dorn was formerly Research Associate at Battelle Memorial Inst., Columbus.

Austenitic Grain Size Control

Austenitic grain size control refers to the control of austenitic grain growth characteristics by means of regulated deoxidation [18, 83]. The effect of small additions of aluminum (one or two pounds per ton) on the growth characteristics of the austenitic grains appears to be a characteristic result of deoxidation with aluminum. On the other hand, steels made with small additions of vanadium, titanium, molybdenum, and other carbide-forming elements also show inhibited austenitic grain growth [25, 45, 59]. Zimmerman, Aborn, and Bain [92] have shown that in the case of steels with small vanadium additions, the inhibition of austenitic grain growth may be the result of low rates of solubility of the more stable carbides. Houdremont, Bennek, and Shrader[45, 59] and others[25, 92] have discussed the subject of the insolubility of carbides, in general, on the properties of steels. These investigations indicate that the mechanism of austenitic growth inhibition is not a direct result of deoxidation in steels with small additions of elements that form difficultly soluble carbides. Aluminum, on the other hand, is not considered to form a very stable carbide and for this reason it has been assumed that its influence on the austenitic grain growth characteristics is a direct result of deoxidation.

McQuaid and Ehn [4] mentioned in their early publication that aluminum additions to experimental heats produced fine austenitic grained steels. Later Epstein and Rawdon [7, 9] showed that commercial fine grained steels were higher in aluminum content than coarse grained steels and demonstrated that additions of 1 lb. of aluminum per ton added in the mold produced fine austenitic grained open-hearth steels. Aplication of these principles to the manufacture of fine austenitic grained steels soon demonstrated that the degree of deoxidation at the time the aluminum was added was an important factor in controlled grain size. Furthermore, different results were obtained for

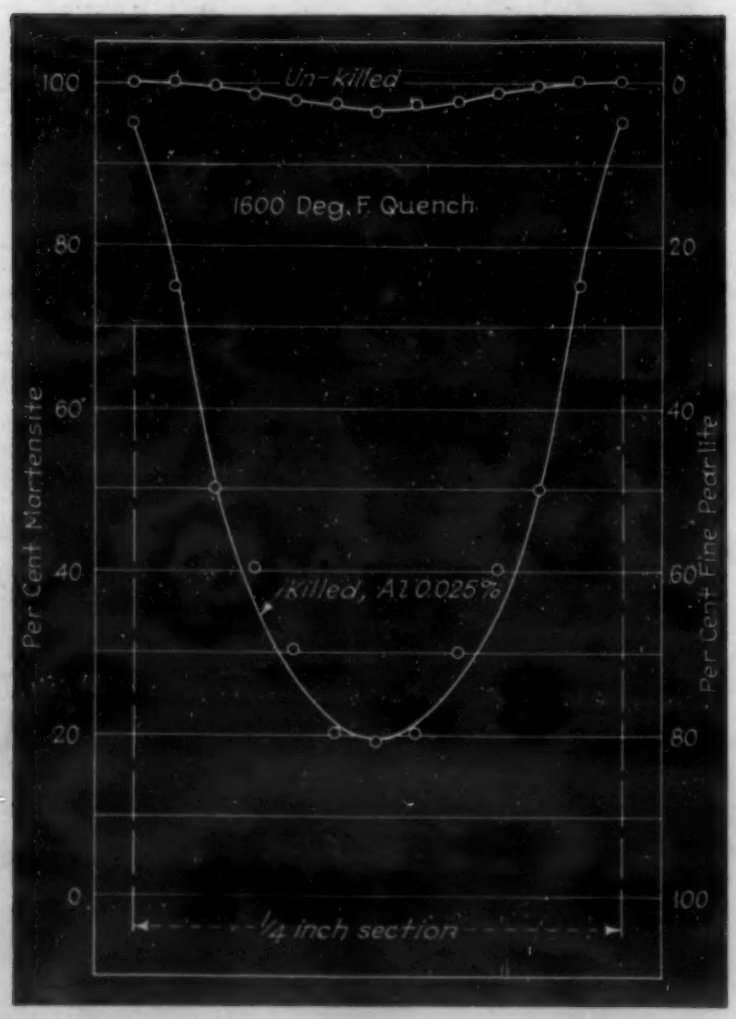
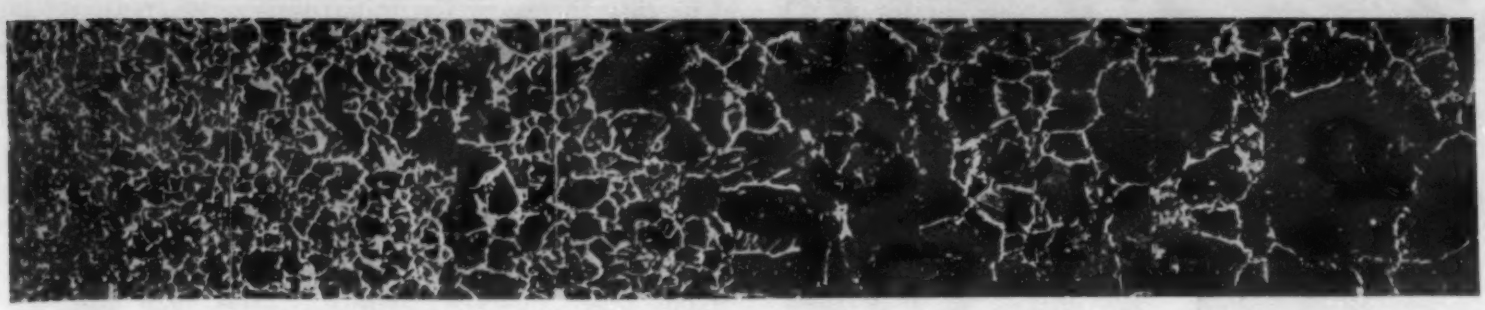


Fig. 23. Curves Illustrating the Difference in the Hardenability of an Aluminum-Killed Steel and an Unkilled Steel. The unkilled steel, which had large austenitic grains, showed almost complete transformation to martensite throughout its cross-section. Only the outer rim of the killed steel, which had fine austenitic grains, transformed completely to martensite. At the center of the quenched cylinder of the killed steel as much as 80 per cent fine pearlite was noted. (Courtesy of E. S. Davenport)

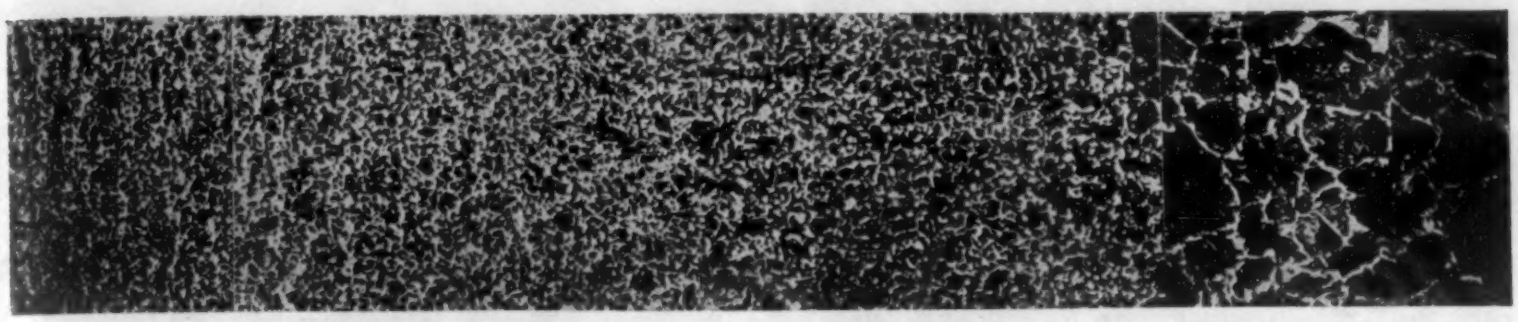
TEMPERATURE—GRAIN SIZE CHART COARSE AND FINE GRAIN S. A. E. 1040

Coarse Grain Steel



1400°F. 1450°F. 1500°F. 1550°F. 1600°F. 1650°F. 1700°F. 1750°F. 1800°F. 1850°F. 1900°F. 1950°F. 2000°F.

Fine Grain Steel



1400°F. 1450°F. 1500°F. 1550°F. 1600°F. 1650°F. 1700°F. 1750°F. 1800°F. 1850°F. 1900°F. 1950°F. 2000°F.

Fig. 24. Assembled Photomicrographs Showing the Effect of Normalizing at Various Temperatures on the Austenitic Grain Size of a Coarse- and a Fine-Grained S. A. E. 1040 Steel. Note that for all normalizing temperatures below 1900 deg. F. the fine-grained steel exhibited almost the same fineness of structure. For this steel the coarsening temperature is slightly below 1900 deg. F. (Courtesy of P. Schane, Jr.)

furnace additions, ladle additions, and additions in the mold [38, 41, 53, 58]. Small quantities of aluminum added in the mold yielded fine grains [38, 41]; ladle additions, in general, had to be larger to yield the same result; much larger furnace additions were usually required to produce fine austenitic grained steels. It was also noted that the effectiveness of ladle additions of aluminum diminished as the time for pouring increased; the last ingot poured had a tendency to have coarse austenitic grains.

The above factors were intensively studied by Epstein, Nead, and Washburn.[38] They demonstrated that good furnace practice need not be changed and, in fact, is highly desirable in order to obtain grain size control. Ladle additions of 1 lb. aluminum per ton to plain carbon steels (0.15 to 0.50 per cent C), containing 0.50 per cent Mn and 0.15 per cent Si or above, yielded fine grained steels. If the silicon content were lower, larger quantities of aluminum were required, and additions of less than one-half pound of aluminum per ton yielded coarse grained steels if the silicon content were less than 0.10 per cent. In the intermediate range, intermediate McQuaid—Ehn grain sizes were not obtained. Inclusion counts showed that moderate amounts of aluminum do not decrease the cleanliness of the steel. In fact, Sims (see *Metal Progress,* Vol. 32, Nov. 1937, p. 674) has demonstrated that the silicate inclusions are deoxidized to form alumina or alumino-silicates, and that this often results in cleaner steels. Other factors, such as the shape and distribution of the sulphides and oxy-sulphides, have been shown by Sims to have a decided influence on the physical properties of aluminum-killed heats. In order to prevent the formation of oxy-sulphides as eutectiferrous constituents in the boundaries of the primary grains, Sims recommends additions of 2 to 4 lbs. of aluminum per ton depending upon the sulphur content of the heat and the furnace practice. In this way, low ductility may be rectified.

Houdremont[73] demonstrated that steels containing 0.025 per cent free aluminum have the maximum austenitic grain coarsening temperatures. As the uncombined aluminum increased, however, the coarsening temperature diminished. This trend did not appear to be associated with the alumina content, which increased uniformly in the range at which this peculiarity was observed.

The effect of aluminum deoxidation on hardenability is shown in Fig. 23. It is generally agreed that controlled austenitic grain growth is a direct result of deoxidation with aluminum, but that the deoxidation is critical and must be closely correlated with the furnace and pouring procedures.

Formation and Growth of Austenitic Grains

Grossman[32] demonstrated that austenitic nuclei are formed at the ferrite-cementite interfaces of the pear

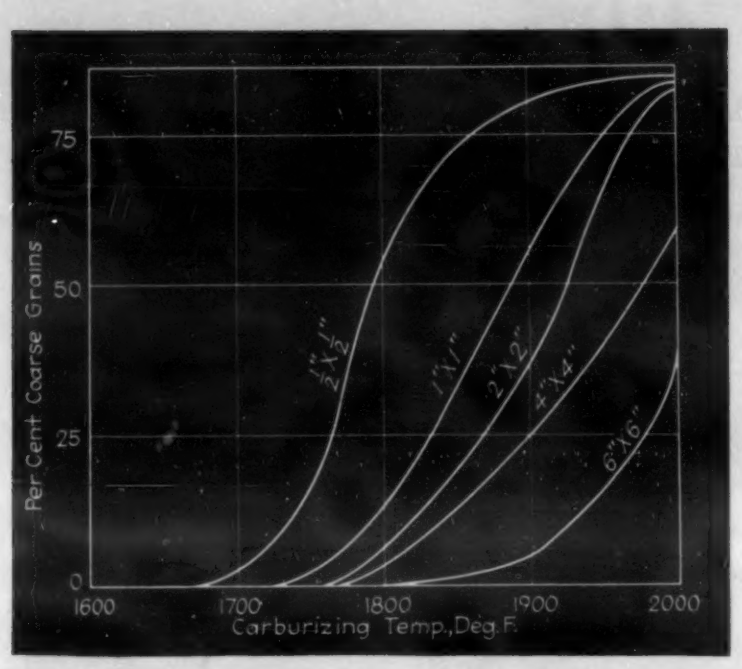


Fig. 25. Curves Illustrating the Influence of Prior Hot Working on the Coarsening Characteristics of the Case of a Carburized Steel. Billets 6 x 6 in. were heated to 2300 deg. F. and given the following successive treatments: Forged from 6 x 6 in. to 4 x 4 in., forged to 2 x 2 in., forged to 1 x 1 in., forged to ½ x ½ in. and cooled in air. The coarsening characteristics were then determined by carburizing for 8 hrs. at a series of temperatures. (Courtesy of M. A. Grossmann)

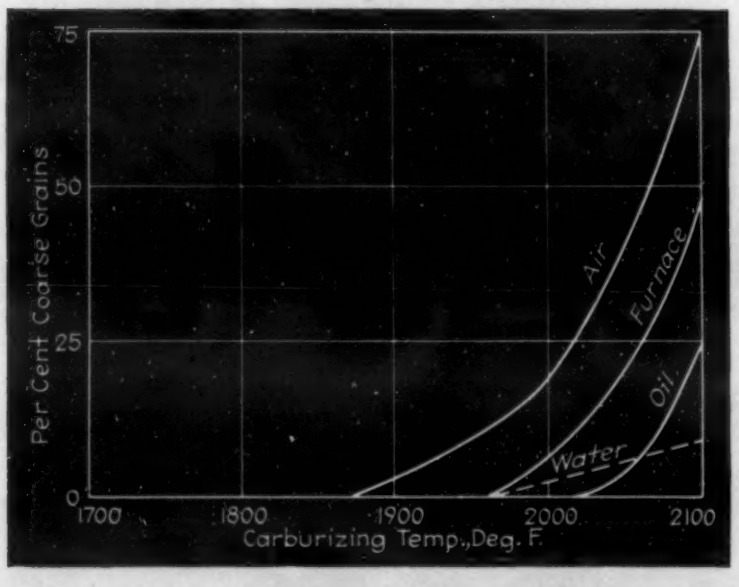


Fig. 26. Curves Illustrating the Influence of Prior Heat Treatment on the Coarsening Characteristics of an S. A. E. 4615 Steel. Samples were heated to 1700 deg. F. and cooled with the furnace, in air, in oil and in water. The percentage of coarse grains after carburizing for 8 hrs. at a series of temperatures is illustrated in the above curves. The only variable is the prior rate of cooling. (Courtesy of M. A. Grossmann)

litic patches when normalized hypoeutectoid steels are heated above the critical range. Such nuclei continue to grow by assimilating adjacent ferrite and cementite until the transformation is complete. The grain size that is developed at this time is called the initial austenitic grain size.

Herty, McBride, and Hough[44] have shown that the initial austenitic grain size of steels that have not had aluminum additions is dependent upon the rate of heating through the critical range. As the rate of heating increases, the initial austenitic grain size becomes larger. If these results be interpreted on the basis of Tammann's theory of nucleation and grain size, it appears that the rate of formation of austenite increases more rapidly, as the temperature increases, than the rate of nucleation. After the initial grain size has been established growth probably continues according to the mechanism discussed by Davey[31]. The grain boundary, which represents a region of transition from the orientation of one grain to the orientation of another grain, migrates by realignment of the boundary atoms adjacent to the stable grain, and a simultaneous disorganization of the external atoms of the less stable grain. When the free surface energy of each grain becomes equal to the same constant value, further growth ceases. Herty, McBride, and Hough[44] have shown that for all practical purposes, the austenitic grain size remains fairly constant after about 20 min. at temperature. Plain carbon steels that have not had aluminum or equivalent alloying additions commonly acquire a No. 1 or No. 2 A.S.T.M. austenitic grain size over a wide range of temperatures. The grain size developed is only slightly greater for higher temperatures, and it increases very leisurely with time at temperature after the first 20 min.

In contrast, however, aluminum-killed steels have vastly different grain growth characteristics. According to Herty, McBride, and Hough[44], the rate of heating through the critical range has very little influence upon the initial size of austenitic grains of aluminum-killed steels. All rates of heating yield small grains, No. 5 to No. 8 A.S.T.M. grain size. There are two possible explanations of this difference between aluminum-killed steels and steels not killed with aluminum; the presence of a large number of points for nucleation outweighs the effect of the rate of heating, or the rate of growth of austenite from nuclei is inhibited. The relative merits of these explanations will be discussed later under theories of inhibited austenitic grain growth.

The austenitic grains of aluminum-killed steels maintain their initial size practically indefinitely at temperatures below the coarsening temperature, according to Grossman[40] and Herty, McBride, and Hough[44]. Growth does not appear to be inhibited,—it actually seems to be blocked or prevented from occuring at all. As the holding temperature is increased, however, a temperature is reached at which grain growth or germination occurs. Numerous investigators have reported that only in the neighborhood of this temperature have duplexed grain sizes been noted[95]. This temperature or range of temperatures is called the coarsening temperature or the coarsening range[40]. Above this range very large austenitic grains are produced. In fact, aluminum-killed steels generally produce larger austenitic grains above their coarsening temperatures than similar unkilled

steels[32, 86, 95]. These characteristics have been studied by Grossman, Bain, and others, and it has been shown that all aluminum-killed steels exhibit these properties and differ essentially with respect to the temperature at which they coarsen. The composite photomicrographs of Fig. 24 illustrate very clearly the difference in grain-coarsening characteristics between coarse and fine grained steels.

According to the above, the important index of austenitic grain growth characteristics of a given steel is its coarsening temperature. Below this temperature, the steel will be fine grained independent of the time of holding, while above this temperature it will behave like a typical coarse austenitic grained steel. It is often desirable to have a coarse grained steel for one operation and a fine grained steel for the following. If then, a steel has a coarsening temperature of 1600 deg. F., it may be made coarse by heat treating at 1650 deg. F. for the first operation and then re-treated at 1550 deg. F. preceding the second operation as described by McQuaid.[47] In order to do this properly, knowledge of the true coarsening temperature of the steel is necessary. Obviously a single test such as the McQuaid-Ehn test[4] cannot yield this information, for in 8 hrs. at 1700 deg. F. the steel in question would be coarse. For this reason, the McQuaid-Ehn carburizing test is too limited to be of great practical importance.

Although there have been several statements in the literature referring to the possible influence of composition on the growth characteristics of the austenitic grains, very little information has been published on this subject. Digges,[94a] however, reported that a series of high purity iron-carbon alloys developed the same grain size (approximately a No. 4 A.S.T.M. grain size) when heated to 1700 deg. F. despite the fact that their compositions varied from 0.23 to 1.21 per cent carbon. The actual temperature and not the displacement above the critical temperature appears to have been the dominant factor in this determination.

Evidence Against "Inherency"

The question concerning "inherent grain size" has been frequently discussed. Obviously the austenitic grain size depends upon time and temperature. In other words, steels do not have "inherent grain sizes." It has also been suggested that the proper expression be considered "inherent grain growth characteristics." This expression is also inaccurate. At present it appears advisable to drop the term "inherent" entirely.

Grain growth characteristics may be altered. It has been shown that cold work influences the austenitic grain growth characteristics of steels. In general, cold working produces finer austenitic grained steels,[32, 90] and Grossmann has attributed this to the finer ferrite grains in cold worked steel. Grossmann[32] has also demonstrated that hot working may reduce the coarsening temperature several hundred deg. F. (see Fig. 25). These examples prove that steels do not have "inherent grain growth characteristics."

Austenitic grain growth characteristics are also influenced by other factors, such as heat treatment (see Figs. 26 and 27) and previous structure. Grossmann [32, 40] and Digges and Jordan[56] found that the greater the degree of subdivision of the cementite the finer the initial austenitic grain size. This appears to be the result of nucleation. Above temperatures at which the carbides completely dissolved, however, the original dispersion of the carbides had little influence on resulting austenitic grain size. Grossmann[32] has correlated to some extent the austenitic grain size with the size of the former ferrite grains. By cooling the same steel from various temperatures from above the upper critical temperature at various rates, a range of structures, from extremely fine to extremely coarse ferrite grains, was obtained. The specimens which had the finer ferritic grains, in general, yielded finer austenitic grains when reheated to certain temperatures. No regular relationship, however, between the initial austenitic grain size and the coarsening temperature was noted. It must be recognized that the tests, as performed, did not eliminate the possible factor that the heat treatments which were used to give a variation in the size of the ferrite grains may have produced changes in the dispersion and solubility of the inclusions.

Dorn and Harder[95] have shown that previous heat treatment has an effect on grain size even if the same structures are initially present. Steels which were held at 1250 deg. F. for about 15 hours, in general, had higher coarsening temperatures than the same steels held at 800 deg. F. for times that yielded approximately the same degree of spheroidization. A second illustration was also presented. One specimen each of a series of steels was heated above the coarsening temperature and quenched. Another specimen of each steel was heated to the same temperature above the coarsening temperature, slowly cooled to just above the upper critical temperature, and quenched. In spite of the fact that both series of specimens were martensitic and had approximately the same size martensitic needles, the specimens quenched directly from the coarsening temperature had higher austenitic grain coarsening temperatures when reheated than the specimens that were slowly cooled and quenched from the lower temperature. In Fig. 28 is shown the effect on austenitic grain size of aging just below the critical temperature.

On the basis of this work, it appears highly probable that some of the effects previously attributed to structure may have been a result of the previous heat treatment that was used to obtain these structures, and is only indirectly attributable to the structures themselves. Nevertheless, reference to Grossmann's[32] investigation on the formation of austenitic grains will

show that the dispersion of cementite, previous to heating, may and very likely does influence the initial austenitic grain size.

Theories of Inhibited Austenitic Grain Growth

There have been a number of attempts to "explain" the austenitic grain growth characteristics of aluminum-killed steels. Although it is generally agreed that the observed facts are a direct consequence of deoxidation with aluminum, both the substance and the mechanism by which inhibited grain growth is accomplished are under discussion. None of the theories appears to explain all of the facts, but some have a more sound theoretical basis and these frequently account for more facts than others. Predictions on the basis of one of the theories[95] have agreed with the effects of heat treatment on the growth characteristics of the austenitic grains in a general way. Although the ability to predict is the most exacting test to which the theory may be subjected, several unexplained facts suggest that an unqualified acceptance of this theory should be postponed until further investigations lead to more reliable conclusions. The attitude taken in this review will be one of criticism of all theories. Emphasis will be placed on the least objectionable theories.

In their first paper on this subject, McQuaid and Ehn[4] suggested that a fine sub-microscopic dispersion of alumina was responsible for the fine austenitic grains found in aluminum-killed steels. This theory has received considerable attention, and it has been assumed by many investigators to be correct. Each investigator, however, has added to the theory his personal interpretation as to what the mechanism may be. It is believed that only a certain fine dispersion of alumina in the steel can effectively inhibit grain growth. This accounts for the differences observed with the deoxidation process, previously mentioned. For example, as the time in the ladle after adding aluminum increases, the submicroscopic alumina particles levigate, and the steel no longer exhibits the well known properties of aluminum-killed steels. Furthermore, if the steel is not well deoxidized before the aluminum is added, the alumina particles that are formed will flux with the excess iron oxide; therefore larger quantities of aluminum are required to produce fine austenitic grained heats of lower silicon content.

Several different mechanisms have been postulated as to how the alumina dispersion is effective in producing fine grains. Some investigators emphasize the nucleation effect of the "refractory" particles. The austenitic grains are assumed to arise from nuclei about each particle of alumina throughout the mass of steel. Since alumina initiates nucleation, aluminum-killed steels result in fine initial austenitic grains. This addition to the dispersion theory does not attempt to account for the fact that these initial grains do not grow, and rather accepts the fact that they are relatively stable.

Practically no point of the above mechanism is acceptable. Grossmann[32] has shown that nucleation takes place at the cementite-ferrite interfaces. Cementite itself promotes nucleation in this reaction. It is highly improbable that the alumina particles are fortuitously located at the ferrite-cementite interfaces. The term "refractory" is a poor connotation for insolubility and low fluxing action. Since heat treatment influences the grain growth characteristics of steel, it is impossible to assume complete insolubility of the inhibiting substance.

Furthermore, the initial austenitic grains are not stable but grow after the coarsening temperature has been exceeded. Evidently some mechanism in addition to nucleation must be effective in preventing grain growth.

Other investigators have assumed that the dispersion of "refractory" particles restrains grain growth. The objection to the term "refractory" has already been discussed. Although it is probable that a dispersion will inhibit or retard grain growth, it would be unlikely to prevent it completely. According to the experiments of Grossmann and others, no growth takes place after the initial grains are formed if the temperature is below the coarsening temperature. Complete blocking and not partial blocking appears to be the experimental fact. Since the conclusions based on the dispersion theory would suggest only a partial blocking of grain growth, this addition to the dispersion theory appears inadequate.

Although more emphasis has been placed upon alumina as the dispersed substance that inhibits grain growth, many investigators[48] have suggested that nitrides and other nonmetallics may act similarly. Derge and Mehl[94] have shown that alumina is active in inhibiting austenitic grain growth, but the highest coarsening temperature they observe in high purity alloys was considerably lower than those commonly found in commercial steels. It is quite probable, therefore, that other aluminum compounds are more active than alumina. This possibility must be given considerable weight in view of a recent publication by Sims and Dahle[98a] illustrating that the growth characteristics of the austenitic grains of steels with aluminum additions depend upon the sulphur content. This may mean that aluminum sulphides or oxysulphides are important grain growth inhibitors. An investigation of this possibility is in progress at Battelle Memorial Institute.

McQuaid[63] recognized a number of the inconsistencies of the various modifications of the dispersion theory, and from time to time suggested several alternative explanations. The theory he has stressed recently is the coalescence theory. McQuaid showed that if steels are held within the lower part of the critical range, those which have an excess of metallic aluminum in solution have greater rates of coalescence

of cementite. It was then assumed that these carbides inhibit austenitic grain growth similarly to the mechanism given by Bain[92] and others[59] for vanadium steels. This theory accounts for the effect of deoxidation on the grain growth characteristics, but fails to account for the fact that large additions of aluminum yield coarse grained steels. Furthermore, the disappearance of vanadium carbide can be directly correlated with the grain growth of vanadium steels[92], while the disappearance of cementite in aluminum-killed steels occurs long before there is any evidence of grain growth. It must be realized that the long hold at the lower critical temperature employed by McQuaid is not representative of the treatment which aluminum-killed steels are given in practice. For these reasons, the coalescence theory has received little support. In fact, it is difficult to understand how the coalescence theory can account for the effect of previous heat treatment on the austenitic grain growth characteristics of steels.

The Adsorption Theory

A fairly sound theory of grain growth has been presented by Benedicks and Löfquist.[67] Since this theory is deserving of considerable attention, it is reproduced in its original form. The paragraphs have been numbered for future reference.

1. The very starting point was the following assumption —which might be designated as the "axiom of grain growth":
2. The migration of a grain boundary depended on the momentary state prevailing in the boundary itself.
3. Thus it did not depend on the conditions of the interior of the grains, otherwise than so far as these influence the properties in (or at) the grain boundaries.
4. The immediate consequences of the axiom was that knowledge of grain growth required a knowledge of the characteristics of the boundary surfaces in general.
5. As for the appearance of boundary surfaces, a strict definition was indispensable, and might be expressed in the following way:
6. The condition for designating two homogeneous-isotropic or anisotropic portions of a substance as physicochemically identical was that the properties scalar as well as vectorial, should be identical in two points inside the two portions.
7. If that were not the case, and the two portions met, a boundary surface would appear. That constituted a limit between the two portions, and these were designated as grains in the case where the substance was anisotropic and possessed a certain solidity.
8. Now, the substance immediately adjacent to the boundary surface, or the surface layer of a grain, generally possessed properties different from those of the interior of the grain. That was an immediate consequence—though hitherto scarcely recognized in the case of an anisotropic solid—of Gibbs fundamental theory of equilibrium. As a matter of fact, Gibbs had explicitly pointed out that "there will be in the vicinity of the surface a certain positive or negative excess of energy, of entropy, and of each of the component substances."
9. In other words, the concentration of a dissolved element must generally be higher or lower in the surface layer (capillary layer or capillary phase) of a grain than in its interior. This fact, theoretically deduced by Gibbs was equivalent to the long-known phenomenon of (positive or negative) adsorption, studied mainly in liquid solutions.
10. It was well known that the solutes which were en-

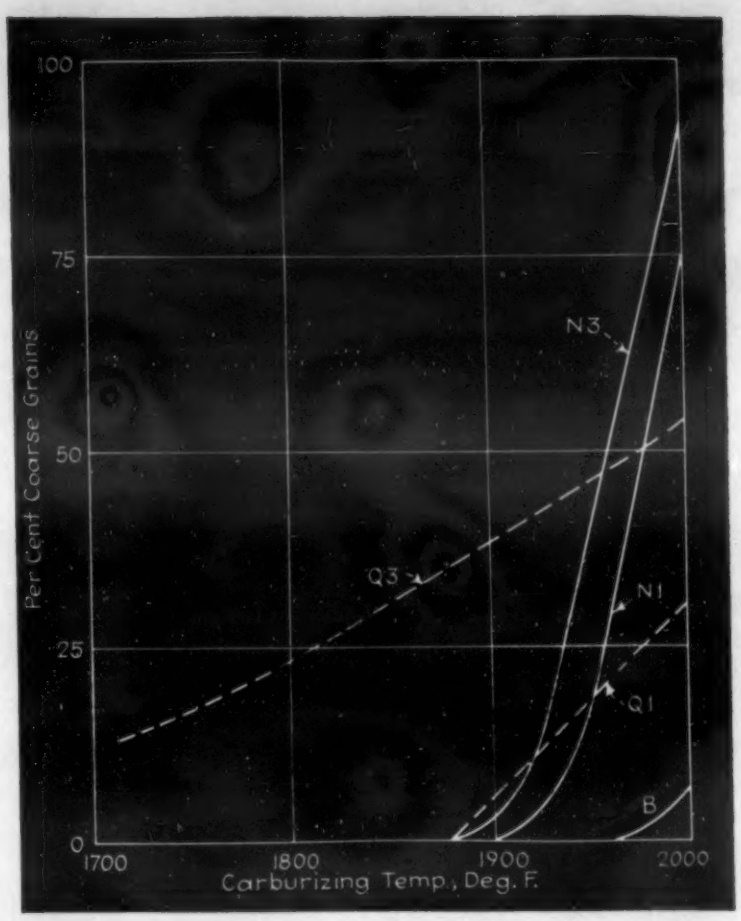
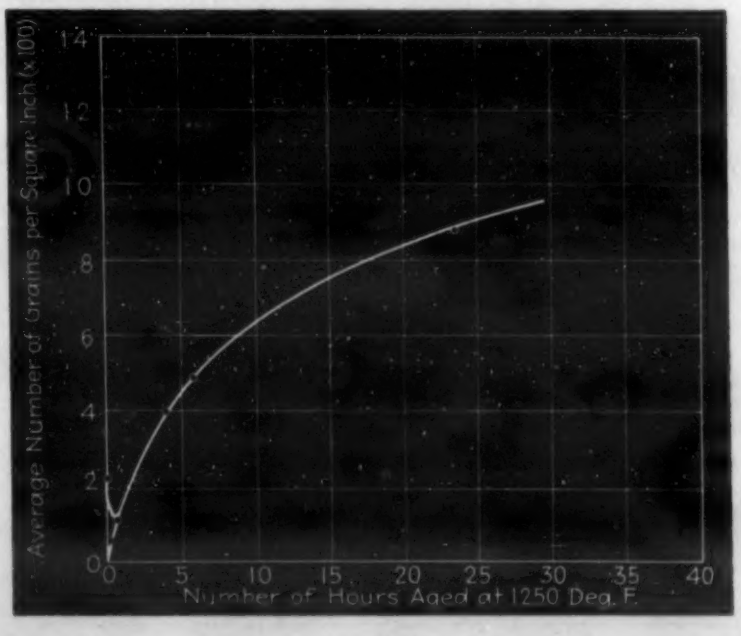


Fig. 27. Curves Illustrating the Influence of Prior Heat Treatment on the Coarsening Characteristics of a Carburizing Steel. B = original billet; N_1 = normalized at 1600 deg. F. once; N_3 = normalized at 1600 deg. F. three successive times; Q_1 = quenched in water from 1600 deg. F. once; Q_3 = quenched in water from 1600 deg. F. three successive times. After these pretreatments the percentage of coarse grains was determined by carburizing for 8 hrs. at a series of temperatures. (Courtesy of M. A. Grossmann)

Fig. 28. Effect of Aging a Hypoeutectoid Steel Just Below the Critical Temperature on the Austenitic Grain Size as Developed in 1 Hr. at 1750 deg. F.



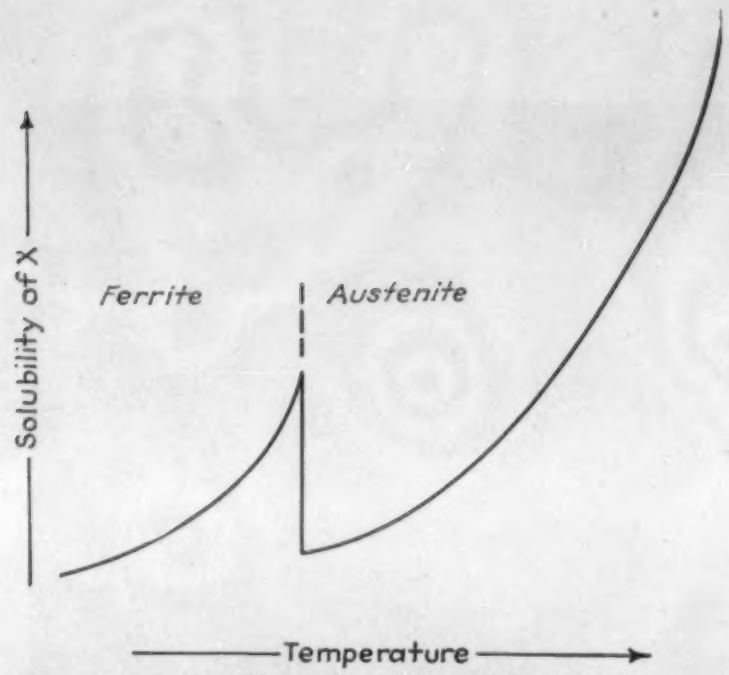


Fig. 29. Curve Illustrating the Relationship between the Solubility of the Unknown Grain Growth Inhibitor X and the Temperature According to the Postulate of Dorn and Harder.

riched in the surface layer were those which considerably lowered the surface tension (Gibbs).

11. To facilitate discussion, substances of this kind, occurring in solid solution, were designated as horophile. On the contrary, elements showing a lower concentration in the surface layer and exerting a negligible (though increasing) influence on the surface tension were designated horophobe.

12. So far, the theory might be said essentially to repose on the solid ground of thermodynamics. The question now arose: What was the influence of interest here, of the higher concentration in the surface layer of a grain, exerted by a horophile substance?

13. Supposing, now, the solute to form with the solvent, or with other substances (impurities) present, a stable hard compound (i.e. known to be hard when existing as a separate phase), then it was natural to conclude that the enrichment would have a certain stiffening action on the surface layer.

14. Accordingly, regarding the boundary between two grains, it was natural to assume that the presence of certain horophile substance might exert a definite stabilizing action on the boundary surface, though not being a "free" one but (according to the above definition) existing between two grains (of the same substance, differently oriented).

15. Such a stabilization of the boundaries would necessarily exert an inhibiting action on the displacement of the boundaries. This was similar to the obviously inhibiting action of a foreign phase precipitated between the grains. In both cases the inhibiting action would prevail, or not, at high temperatures, depending on the "refractivity" or insolubility of the boundary layer.

16. Thus, it was by no means necessary for an inhibiting action on grain growth that a foreign substance should form a distinct phase, acting as a barrier between two grains; a barrier action would result even from a capillary enrichment, or adsorption in the boundary layer.

17. Such an enrichment would naturally cause a decrease of the cohesive forces between two adjacent grains, and might thus influence the mechanical properties of the material.

18. Detailed treatment required the application of principles of energy, or of the action of surface tension. As a matter of fact, in a pure substance—which, as was shown, was very liable to form large grains—the driving force of grain growth might be considered the tendency of the surface energy, as defined by the surface tension multiplied by the total boundary area, to become a minimum. If a horophile substance were present, lowering the surface tension, the tendency for grain growth would evidently decrease, simply from a general energy point of view. Besides this, the more mechanical stabilizing effect treated above might considerably increase the stability of the grains.

It is refreshing to note that Benedicks and Lötquist actually pointed to the grain boundary as the source of inhibited austenitic grain growth. This appears to be the most logical approach so far discussed. Undoubtedly the grain boundary adsorption theory is pertinent to grain growth characteristics. It does not, however, account for a usually definite coarsening temperature. Contrary to paragraph 15 of the adsorption theory, the adsorbed material must be of the same phase as the grain; therefore it forms a solid solution of higher concentration at the grain boundary, and higher temperatures cannot cause the inhibiting substance to dissolve since it is already in solution. This difficulty was avoided by the assumptions made in paragraph 16. Nevertheless, this theory accounts very well for most of the observed phenomena. It is doubtful, however, that the theory will account for the effects of previous heat treatment. It appears that if a definite quantity of the inhibiting material is present, the same coarsening temperature will result, independent of previous treatment.

The Film Theory

Before McQuaid postulated the coalescence theory, he suggested that inhibited austenitic grain growth may be due to an inhibiting film about the grains.[48] Shane[50] showed that such a theory agrees with many of the facts, especially that of a definite coarsening temperature. Dorn and Harder[95] extended this theory to account for the formation of this film and the effect of previous treatment on the growth characteristics of austenitic grains. They assumed that some product of aluminum (which they referred to as X, the unknown inhibitor) is slightly soluble in ferrite and austenite. As the temperature increases, the solubility of X in ferrite increases up to the transformation temperature. At this point, the solubility in ferrite was assumed to be greater than that in austenite. In the austenite range the solubility increases with the temperature. The solubility curve of X is believed to be of the type shown in Fig. 29. Preliminary tests of this hypothesis indicated that (1) the excess X in solution in ferrite over the saturation value in austenite precipitated in the austenitic grain boundaries and prevented further growth; (2) at the coarsening temperature, X dissolved at its thinnest points, causing duplexed grain sizes; (3) above the coarsening temperature, complete solution of X took place and uniformly large grains were produced. Conclusions on the basis of the solubility of X were found to be in agreement with other experimental relationships[95] between the previous heat treatment and the growth characteristics of austenitic grains.

Although the film theory appears to account for more of the facts than any of the previously mentioned theories, there is at least one factor with which it is not wholly compatible. A film of sufficient thickness to prevent growth may also be expected to pre-

vent the diffusion of carbon. Experiments have shown that although the rate of diffusion of carbon in aluminum-killed steels is less than in unkilled steels, diffusion does take place. It appears necessary to assume, at present, that the film prevents grain growth but that inhibited diffusion of carbon through this film is possible.

Although much progress has been made since McQuaid and Ehn pointed out that the "quality" of steels deoxidized with aluminum differs from the "quality" of those not deoxidized with aluminum, there are many important commercial and theoretical aspects that still require explanation. Our knowledge on austenitic grain size is yet in its infancy.

Acknowledgements

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All references published in March issue.

New Alloys in the Optical Industry

BY W. J. WRIGHTON AND TRACY C. JARRETT

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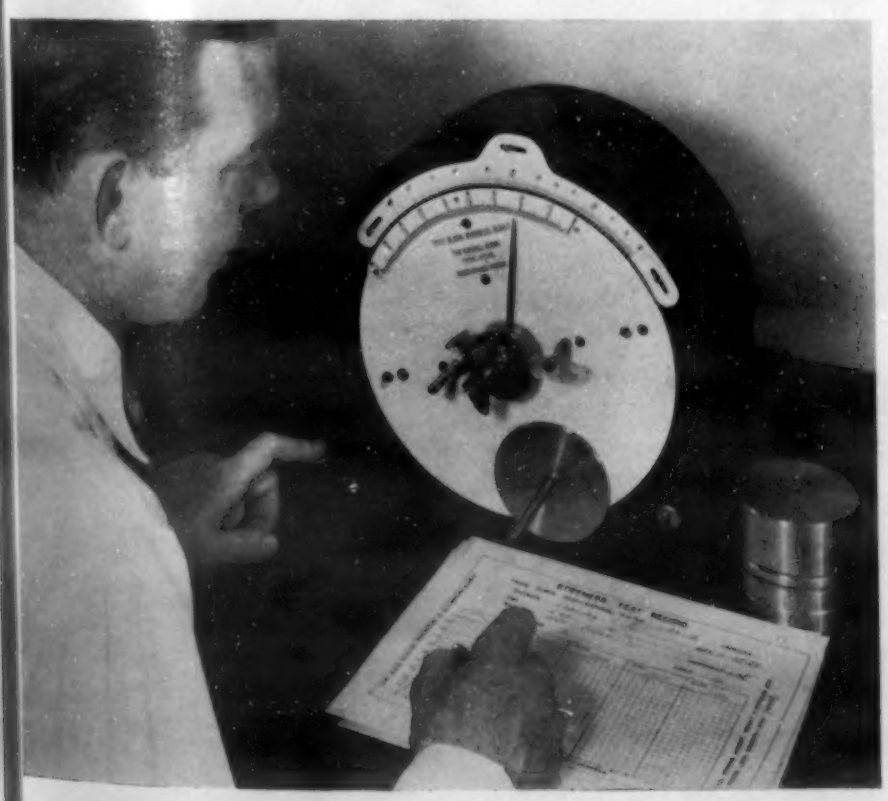


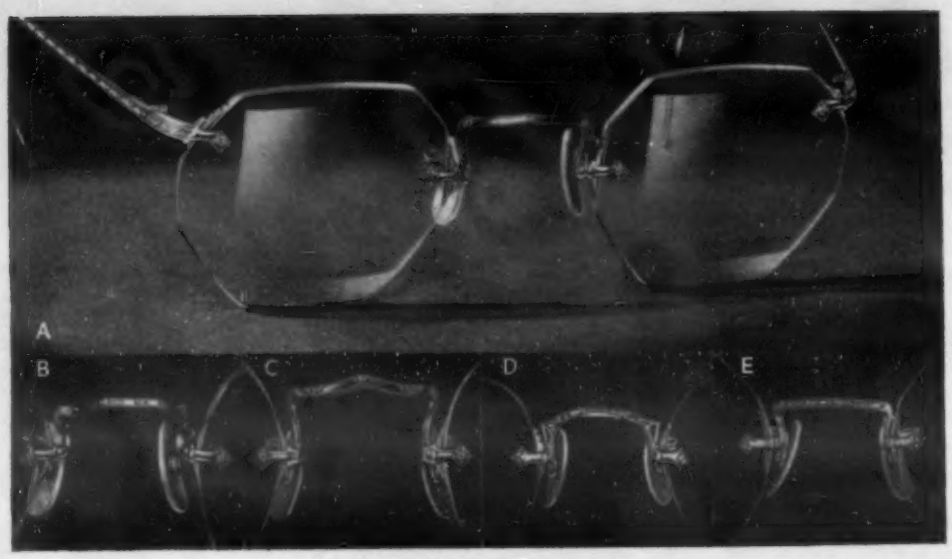
The Stiffness Tester for Determining the Properties of Metals for Spectacle Bridges.

IN THE OPTICAL FIELD, like any other, as a general rule there is always a desire to improve the products manufactured. In addition, competition among manufacturers inevitably brings about improvement, and after a time the purchasing public begins to demand further improvements. The fact that new alloys are on the market today helps to bring about these changes. In the early days of spectacles, the public accepted the product that the manufacturer was able to make and the demands at that time were not great.

The reverse is now true since the manufacturer has to produce what the public requires. In the early days of spectacle making style changes played little part but today they are very important. Functional features, such as flexibility, stiffness and acid resistance also play an important part. There has always been a need for a great many of these physical and chemical properties, but it has only been within recent years that these demands have been met.

First Made of Steel

Spectacles were first made from steel and, as they



Six Different Styles of Bridges for Spectacles.

were largely hand made, it was possible to produce from this material optical goods which satisfied the demand at that time. When the demand for better appearance and increased production came about, gold-filled materials were introduced, replacing steel. Gold-filled bronzes and nickel silvers were the materials available at that time. ("Gold Filled Plate and Wire in the Optical Industry" METALS AND ALLOYS, May, 1938, Vol. IX, p. 127-130.) The same base metal could not be used for everything, for in some cases stiffness was desired, and in others flexibility.

One of the standard methods for producing gold-filled bridges was to use a completely gold covered wire, swaging and bending to the desired shape. This produced satisfactory results but limited the styles available. Style changes became more frequent. Along with this the blanked out bridge which permitted more style changes was brought into use. The blanked out bridge was made from flat stock, with gold on both sides which left the edges of the bridge exposed after blanking. Even though the edges of the bridge were well protected by an electro plate, this plate had a tendency to wear off. And particularly if a bronze was used, facial acids would attack the base metals and tend to destroy the bridge.

Acid Retarding Metals

This brought into the picture acid retarding metals such as pure nickel, as they withstood the facial acids under adverse climate conditions better than the red bronzes and nickel silvers. This was a step forward in the optical field, although some of the desired physical properties were not met.

With the introduction and use of the nickel-chromium alloys, another stride was made in the optical field. Here again, some of the physical properties were lost, but their ability to withstand facial acids is excellent. When acid resisting alloys took their place in the optical field, stainless steels also were tried, but are not used for gold-filled material because they lack good physical properties. All the above alloys mentioned are hard on tools and dies, in spite of the fact that good tool steels are available.

Beryllium-Copper Introduced

Along with the new alloys came beryllium-copper. This alloy, although it does not have the corrosion resistance of the nickel-chromium alloys, it does have excellent physical properties. It is possible to use this completely gold covered alloy soft, and then by heat treating obtain the desired physical properties. There are other alloys that were tried but proved to be of little value.

In making optical bridges, style plays a very important part as to the base metal to be used. In rimless bridges, such as are shown in one of the illustrations, several important features must be met: First, the bridge should have good flexibility. Second, it should have good stiffness, and the ability to withstand plier adjustments, to which they are submitted when fitting them to a person's face.

There are six different centers, each one having different properties. To maintain a certain standard and to obtain the desired properties, different base metals must be used. A satisfactory method for determining the properties for spectacle bridges is the stiffness tester as illustrated. This test gives the height of maximum bending moment or stiffness and the degree of set that a bridge will have when a constant load is applied to a definite angle. Table 1 shows the physical properties of various styles. In Table 2, the effect of different base metals upon one style is brought out.

The use of new alloys applies not only to bridge but to other spectacle parts.

Table 1—Physical Properties of Various Styles

No.	Base metal	Style of bridge	Max. bending moment	Degrees set	Characteristics
A	Beryllium copper..	Reduced	82	14	Flexible
B	Beryllium copper..	Reduced	65	13	Flexible
C	Nickel-silver	Reduced	70	20	Semi-flexible
D	Nickel-chromium ..	Blanked	72	21	Semi-flexible
E	Beryllium copper..	Reduced	72	20	Semi-flexible

Different styles require different base metal. Bridges A, B, and E are made of 2.25 per cent Be-Cu. Bridge C is 10 per cent nickel-silver (Ni 10, Cu 67, Zn 23%). Bridge D is Inconel, approximately Ni 80, Cr 12 and Fe 8 per cent. The reduction of area of each bridge is as follows: A 8, B 1?, C 33, D 54, and E 22 per cent.

Table 2—Effect of Different Base Metals on One Style

No.	Base metal	Style of bridge	Max. bending moment	Degrees set	Characteristics
A	Beryllium copper..	Reduced	82	14	Flexible
A	Nickel	Reduced	88	26	Stiff
A	Nickel-silver	Reduced	65	20	Soft
A	Copper-bronze	Reduced	67	20	Soft

The copper bronze has the following composition: Cu 90, Sn 5, Zn 5 per cent. The reduction of area of these bridges is: Be-Cu, 8; Ni, 27; nickel-silver, 27 and copper bronze, 27 per cent. These are on the wire as it started into its reducing and swaging machines. Final degree of cold work is not available because these bridges are submitted to several different swagings, bendings and engravings during the processing.

Trends in Alloy and Electric Steel Output

by Edwin F. Cone

We believe that analyses of reliable production statistics are able to show significant trends. In this article an attempt has been made to reveal the progress in the demand for alloy steels in general, and the stainless steels in particular, based on the quantity made, and hence to gage the importance of this branch of the steel industry. Alloy steel production is so intimately tied up with the electric steel industry that an analysis of the trend in the output of steel from electric furnaces is included.

Some interesting and significant trends are pointed out. The basis of the analysis is different from that used in former articles of this nature.—E.F.C.

STATISTICS ARE VALUABLE as indicating a trend. Unless they are fairly complete, however, their significance is impaired, with deductions based on them likely to be misleading.

Metallurgical engineers, producers and users are keenly interested in the progress from year to year of the American alloy steel industry and of electric steel. To trace this progress through the entire steel industry, complete data of the production of not only steel ingots, but also of steel castings, are necessary.

Unfortunately statistics are not available of the output of steel foundries in this country. The American Iron and Steel Institute, since 1933, collects steel casting production of only those companies making ingots —the official data do not represent all castings produced. The Steel Founders Society of America assembles only the statistics of its members—hence its data are not complete. The Alloy Casting Association, representing most producers of corrosion and heat-resisting steels (the stainless steels) collect production data, but do not publish them.

During the last two or three years, we have published[1] an attempted estimate of the steel foundry production and also an estimate of the trend in alloy steel. To the statistics of the American Iron and Steel Institute (which give accurate data on ingots with castings output of ingot companies) we added estimates of foundry output. Because the foundry data are difficult to obtain and interpret, we have decided to base the present survey of progress in the alloy steel and the electric steel industries on only the reliable data of the Institute.

There is, therefore, here presented, an analysis of the trend in the American alloy steel industry with data and comments in the electric steel industry, based on the original statistics of the Institute for 1938 and previous years. Omission of the major portion of the steel foundry output does not seriously affect a judgment of the trends.

Total Alloy Steel

In Table 1 there are assembled data of the production of steel ingots and castings, with the total alloy steel output and its relation to the total steel and so on. To, and including 1933, the data include both ingot and casting totals, but from 1934 on the statistics (in italics) give only the steel castings as produced by ingot making companies, distinguished by publication in the table as italics.

The 1938 output of alloy steel ingots and castings was 1,434,801 gross tons, or the smallest since 1933 —due of course to the "recession in business." Table

One Hundred Tons of Molten Pig Iron Being Poured into an Open-Hearth Furnace for Conversion into Steel. Large quantities of alloy steels are made in this type of furnace. (Courtesy: Carnegie-Illinois Steel Corp.)

Table 1.—Production of Alloy Steel Ingots and Castings in United States—Gross Tons.

	Total Alloy Steel	Total Steel	Per Cent Alloy Steel	Tons of Total Steel to 1 Ton of Alloy Steel
1909	181,980	23,955,021	0.75	131.6
1910	567,819	26,094,919	2.17	45.9
1911	481,459	23,676,106	2.03	49.1
1912	792,501	31,251,303	2.53	39.4
1913	714,357	31,300,874	2.28	43.8
1914	646,953	23,573,030	2.75	36.3
1915	1,021,147	32,151,036	3.17	31.4
1916	1,362,615	42,773,680	3.18	31.3
1917	1,644,335	45,060,607	3.65	27.4
1918	1,787,852	44,462,432	4.02	24.8
1919	1,481,188	34,671,232	4.27	23.4
1920	1,660,292	42,132,934	3.94	25.3
1921	809,548	19,783,797	4.10	24.3
1922	1,673,496	35,602,926	4.70	21.3
1923	2,106,489	44,943,696	4.70	21.3
1924	2,026,409	37,931,939	5.35	18.7
1925	2,432,973	45,393,524	5.36	18.6
1926	2,463,414	48,293,763	5.10	19.6
1927	2,531,748	44,935,185	5.59	17.7
1928	3,214,909	51,544,180	6.23	16.0
1929	3,957,207	56,433,473	7.01	14.2
1930	2,443,311	40,699,483	6.00	16.7
1931	1,455,913	25,945,501	5.61	17.8
1932	798,604	13,681,162	5.84	17.1
1933	1,547,183	23,232,347	6.65	15.0
1934	*1,612,275*	*26,055,289*	*6.17*	*16.1*
1935	*2,119,658*	*34,092,594*	*6.21*	*16.0*
1936	*2,883,622*	*47,767,856*	*6.04*	*16.5*
1937	*3,032,626*	*50,568,701*	*5.98*	*16.6*
1938	*1,476,348*	*28,349,991*	*5.20*	*18.1*

Table 2.—Production of Alloy Steels made in Electric Furnaces in United States—Gross Tons.

	Electric	Total Alloy Steel	Per Cent of Electric Alloy Steel of Total Alloy
1909	.....	181,980	....
1910	608	567,819	0.11
1911	.....	481,459	....
1912	9,619	792,501	1.21
1913	11,264	714,357	1.57
1914	9,344	646,953	1.44
1915	27,944	1,021,147	2.73
1916	71,129	1,362,615	5.22
1917	130,578	1,644,335	7.94
1918	290,961	1,787,852	16.26
1919	181,632	1,481,188	12.26
1920	245,572	1,660,292	14.78
1921	63,246	809,548	7.81
1922	125,419	1,673,496	7.49
1923	194,976	2,106,489	9.25
1924	188,563	2,026,409	9.30
1925	293,780	2,432,973	12.00
1926	306,811	2,463,414	12.45
1927	343,517	2,531,748	13.57
1928	433,096	3,214,909	13.47
1929	510,030	3,957,207	12.87
1930	300,520	2,443,311	12.51
1931	232,113	1,455,913	15.94
1932	140,877	798,604	17.65
1933	296,210	1,547,183	19.11
1934	*299,236*	*1,612,275*	*18.5*
1935	*412,563*	*2,119,658*	*19.4*
1936	*527,762*	*2,883,622*	*18.3*
1937	*600,550*	*3,032,626*	*19.8*
1938	*332,475*	*1,476,348*	*22.5*

Table 3.—Production of Stainless Steel Ingots—Gross Tons

	"18 & 8"	12 to 14% Cr	16 to 18% Cr	All Other	Totals
1934	19,777	8,902	8,077	13,161	49,917
1935	24,711	13,232	12,387	14,917	65,697
1936	39,107	23,598	21,288	6,973	90,966
1937	59,118	30,302	24,598	25,820	139,838
1938	38,508	11,990	14,691	20,484	85,673

Table 4.—Proportion of Stainless Steel of Total Alloy Steel—Gross Tons

	Stainless Ingots and Castings	Total Alloy Steel	Per Cent Stainless
1934	49,917	1,612,275	3.10
1935	65,697	2,119,658	3.09
1936	90,966	2,883,622	3.16
1937	139,834	3,032,626	4.61
1938	85,673	1,476,348	5.82

1 shows that the percentage of alloy steel of the total steel output fell last year to 5.20 per cent after standing a number of years at 6 per cent or over. In tons of total steel to 1 ton of alloy steel the record for 1938 advanced to 18.1 tons after remaining for a number of years fairly steadily at 16 to 17 tons. This is of course a loss.

Electric Alloy Steels

Alloy steels are made in both open-hearth and electric furnaces. It is interesting to analyze the trend in the proportion made by electric melting. Table 2 is formulated to show this.

The percentage of electric alloy steel of the total alloy steel produced has increased quite decidedly in the last two years. At 22.5 per cent for 1938 the highest percentage recorded has been reached. Even at 19.8 per cent in 1937, a new record at that time was made. These facts lend substantiation to the contention of some that electric furnaces are likely to make an increasing proportion of the alloy steels.

Stainless Steel Production

The interest in the stainless steels, or the corrosion and heat-resisting steels, is mounting each year. Only since 1934 have reliable data on the country's output been available. The Institute now publishes these data each year and they are reproduced in Table 3.

Total production of these steels in 1938 at 85,673 gross tons was the smallest since 1935. As indicated, the "18 and 8" type predominates. In the last three years, it has represented 42 to 44 per cent of the total. The 12 to 14 per cent Cr brand ranks second.

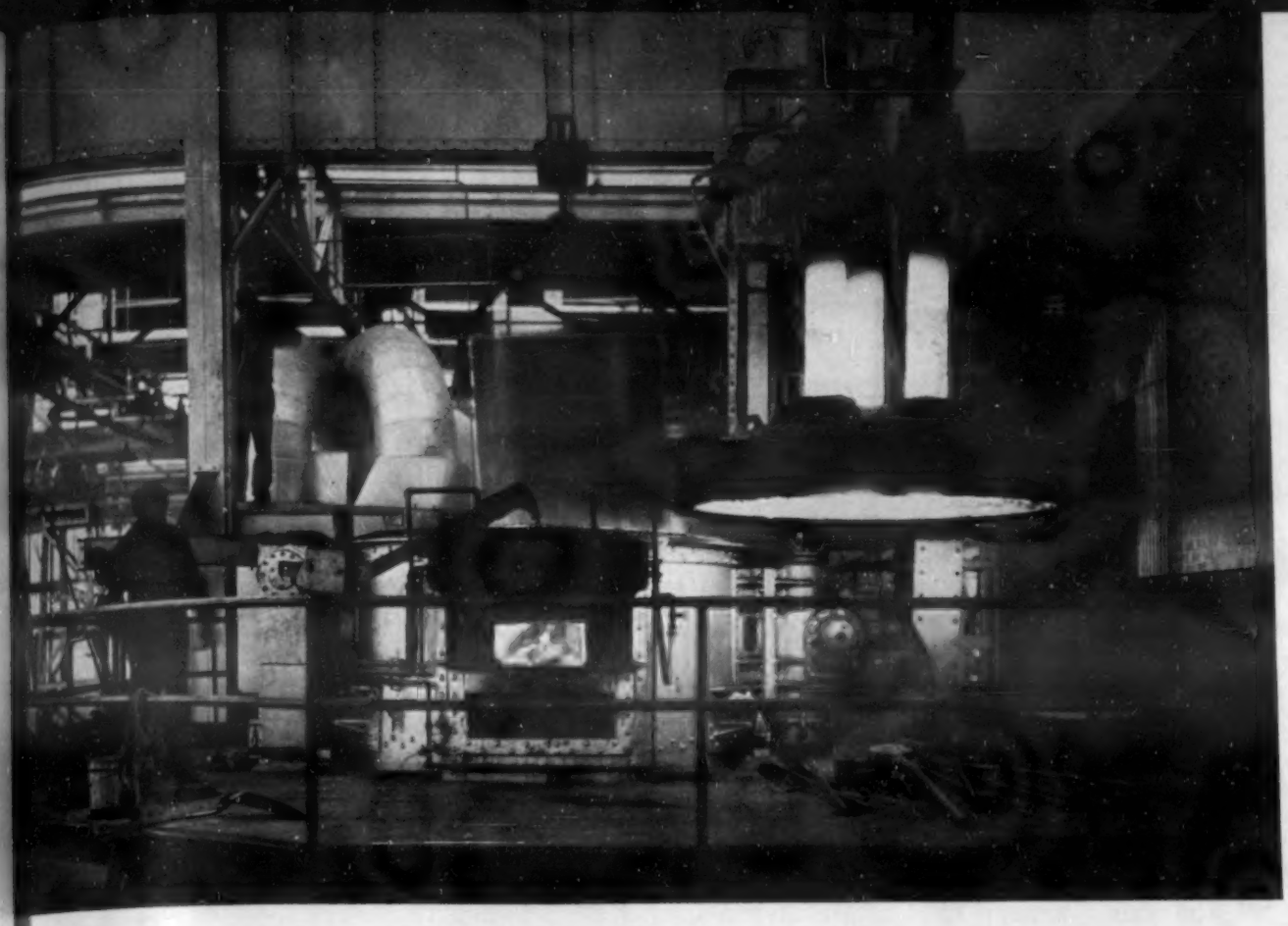
An interesting sidelight is the relation of the stainless steel totals to the total alloy steel output. Table 4 reveals this. It shows that, whereas in 1934, 1935 and 1936, the percentage of stainless steel averaged around 3 per cent, it has advanced markedly in the last two years. A high of 5.82 per cent of 1938 reflects the increasing demand and use of these steels.

Electric Steel Production

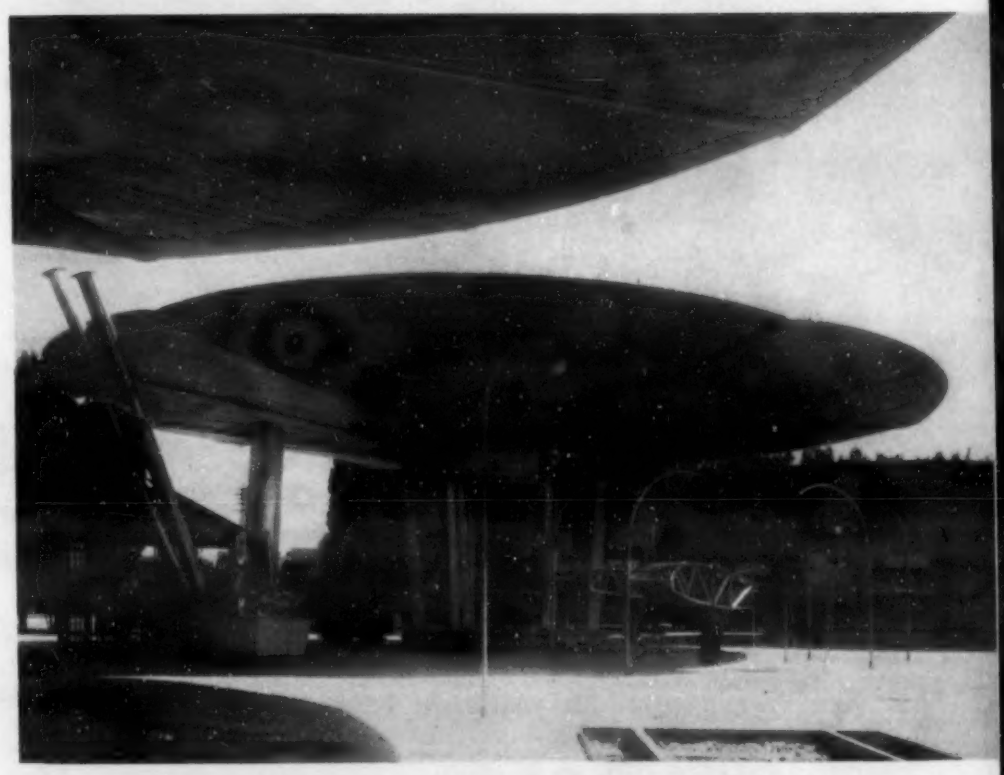
The production of electric steel in the United States in the last 10 years has been fairly steady with some gains. Table 5, based on the Institute's data,

Table 5.—Electric Steel Output—Gross Tons

Year	Total Electric	Total Steel	Per Cent Electric
1920	505,687	42,132,934	1.20
1926	650,723	48,293,763	1.35
1929	951,431	56,433,473	1.68
1933	421,203	23,232,347	1.81
1934	*361,296*	*26,055,289*	*1.38*
1935	*541,492*	*34,092,594*	*1.58*
1936	*772,455*	*47,767,856*	*1.62*
1937	*845,537*	*50,568,701*	*1.67*
1938	*505,024*	*28,349,991*	*1.78*



A Top-Charge Swindell-Dressler Arc Electric Furnace. All stainless steels are made in arc or Ajax-Northrup high frequency furnaces as well as other high grade alloy steels. (Courtesy: Ford Motor Co. of Canada)



Giant Umbrellas, Fabricated of Stainless Steel, Shelter the Exhibit of the Edward G. Budd Mfg. Co. at the New York World's Fair, 1939. They are 60 ft. in diameter, and four in number.

tells the story of development. From 1934 on, it must be remembered, the data do not include electric steel castings, the output of which is not small. These data are distinguished by italics in the table, because the other data include all steel castings made.

The percentage of electric steel of the total steel produced has, since 1920 and the earlier years, been characterized by a steady growth. The years of depression have not seriously cut down the percentages. From 1933 to 1938 inclusive, the percentage of electric of total steel has averaged, with one exception, from 1.58 to 1.81 per cent, with the 1938 figure of 1.78 per cent, close to the high. In 1920 the figure was only 1.20 per cent with an advance to 1.35 per cent in 1926 and to 1.68 per cent in 1929. All signs point to an even greater increase in coming years.

Summary

In recapitulation of the data presented, the following outstanding features may be cited:

The 1938 output of alloy steels made the poorest showing in some years as judged by the data in Table 1. The trend, however, has been and will be steadily upward.

Percentage of alloy steels made in electric furnaces in 1938 reached a new high.

In the stainless steels, the total 1938 output was the smallest in 3 years.—18 and 8 continues to lead the procession.

Of the total alloy steel, the percentage of stainless steels in 1938 at 5.82 per cent was the highest recorded.

Total electric steel production in 1938 was 1.78 per cent of total steel, the second highest percentage recorded.

References

[1] "Trends in Stainless Steel Production," December, 1935, Vol. 6, page 347.

"Trends in American Alloy Steel Production," November, 1935, Vol. 6, page 329.

"Trends in American Alloy Steel Production," June, 1936, Vol. 7, page 162.

"Production Trends of American Alloy Steel and Electric Steel," July, 1937, Vol. 8, page 191.

"Production Trends of American Alloy Steel and Electric Steel," August, 1938, Vol. 9, page 207.

EDITORIALS (*Continued from page 227*)

topics in the same way, if publication is withheld.

Papers committees and editors can, however, keep down the number and the bulk of unimportant publications by refusing those articles that are blatant propaganda written for sales purposes alone, and by drastic stripping of the bulky shells from those whose kernels are tiny.—H. W. G.

Scrap Iron and Steel for Japan

In view of the discussion, heard so often now, in the daily press, regarding shipments of "war materials" by Americans to Japan and their effect, we looked into the official statistics for some light. The result is rather astounding—particularly as regards scrap iron and steel.

The large movement of scrap to Japan began in 1933 when we sent 547,500 gross tons to that country out of total exports of 773,400 tons—or over 70 per cent. Since then the total exports and the sales to Japan have been increasing tremendously right down to the present time—larger in some years than others, but very large.

Without going into details, the data show that in 1937 we sent to Japan over 1,904,000 tons of scrap, including tin plate scrap, out of a total of 4,092,500 tons or 46.5 per cent. Italy took 838,200 tons of this total. This has been the peak year in tonnage. From 1933 to Apr. 1, 1939 (6 yrs. and 3 months) the United States has exported about 14,514,000 tons of scrap iron and steel of which 7,772,400 tons or 53.5 per cent has been taken by Japan. The two other large buyers have been Italy and Great Britain.

Just to emphasize the picture: In copper—during 1936 and 1937 we exported to Japan 225,541,000 pounds of refined copper and 16,194,000 pounds of scrap copper—by no means insignificant!

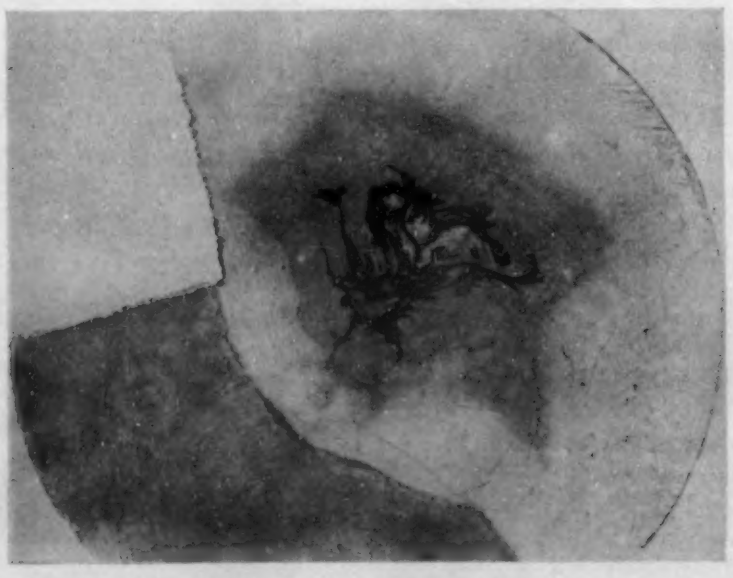
The natural thought based on these facts is—what would have been the status of Japan if these sales had been shut off. Over 7,772,400 tons of scrap steel means almost that much refined steel as ingots—over 2,500,000 tons larger than Japan's 1936 ingot steel output which was a record. There are no other sources of supplies of that magnitude other than the American. Those who argue that, for a substantial profit, Americans have contributed potently to the destruction of friendly countries, China in particular, and are at the same time assisting in the creation of a "Frankenstein," are correct—with results no man can foresee.

Whether or not we are partly exhausting our supplies of scrap steel or copper, the situation should be studied by Washington and in the interests of peace and humanity, possibly some limit set on such exports or an embargo to nations at war, actual or undeclared.—E.F.C.

A FEW CHUCKLES

"A Chinese Drago~"

To the Editor: I have noticed a few curiosities in microstructures in your column titled, "A Few Chuckles," and I thought that you might be interested in a "Chinese



dragon" entrapped in such a structure. This was found at the junction of the parent metal and the weld metal in a 2½-in. round and is composed of iron oxide and flux.

Algoma Steel Corp. Ltd.
Sault Ste. Marie
Ontario, Canada

F. J. McMULKIN,
Metallographer.

A Furnace is a Furnace!

At the large exhibition of one of the leading American steel companies at the New York World's Fair, 1939, one of the demonstrators was asked to explain the function of the blast furnace as represented by the miniature reproduction or diorama. This was painstakingly done and when completed, the inquirer said: "Have any of these furnaces been installed yet in houses for heating?!!"—E. F. C

Another Chuckle on Us!

Coming home tired and weary, I like to pick up METALS AND ALLOYS and, if the new issue has come, I always look first of all for any new metallurgical chuckles. In the December, 1938, issue, I found one which appealed to me as a good "Chuckle" would, but it is not so designated. If you will look at Page MA 728, left-hand column, line 15 from the bottom, you will read as follows: "The austenite not transformed in this transformation remains unchanged."

V. N. KRIVOBOK.

What's in a Name?

To the Editor: For a chuckle—see Caption of Fig. 1, p. 78, *The Iron Age,* May 18, 1939—"*Amorphous* Cementite"

M. C. FETZER.

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